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Volume 195 Part 4

Established 1869
Incorporated by Royal Charter 1899

Report of the Annual General Meeting 377

Adherence of oxide films on metals
R. F. Tylecote 380

Physical metallurgy of 12%Cr steels
K. J. Irvine et al. 386

Under-hearth cooling for blast-furnaces
E. M. Summers 405

Effect of aluminium and nitrogen on the occurrence of intergranular fracture in steel castings
B. C. Woodfine and A. G. Quarrell 409

Practical estimation of composition balance and ferrite content in stainless steels
L. Pryce and K. W. Andrews 415

Radiochemical investigation of the combustion method for determination of sulphur in cast iron
R. C. Rooney and F. Scott 417

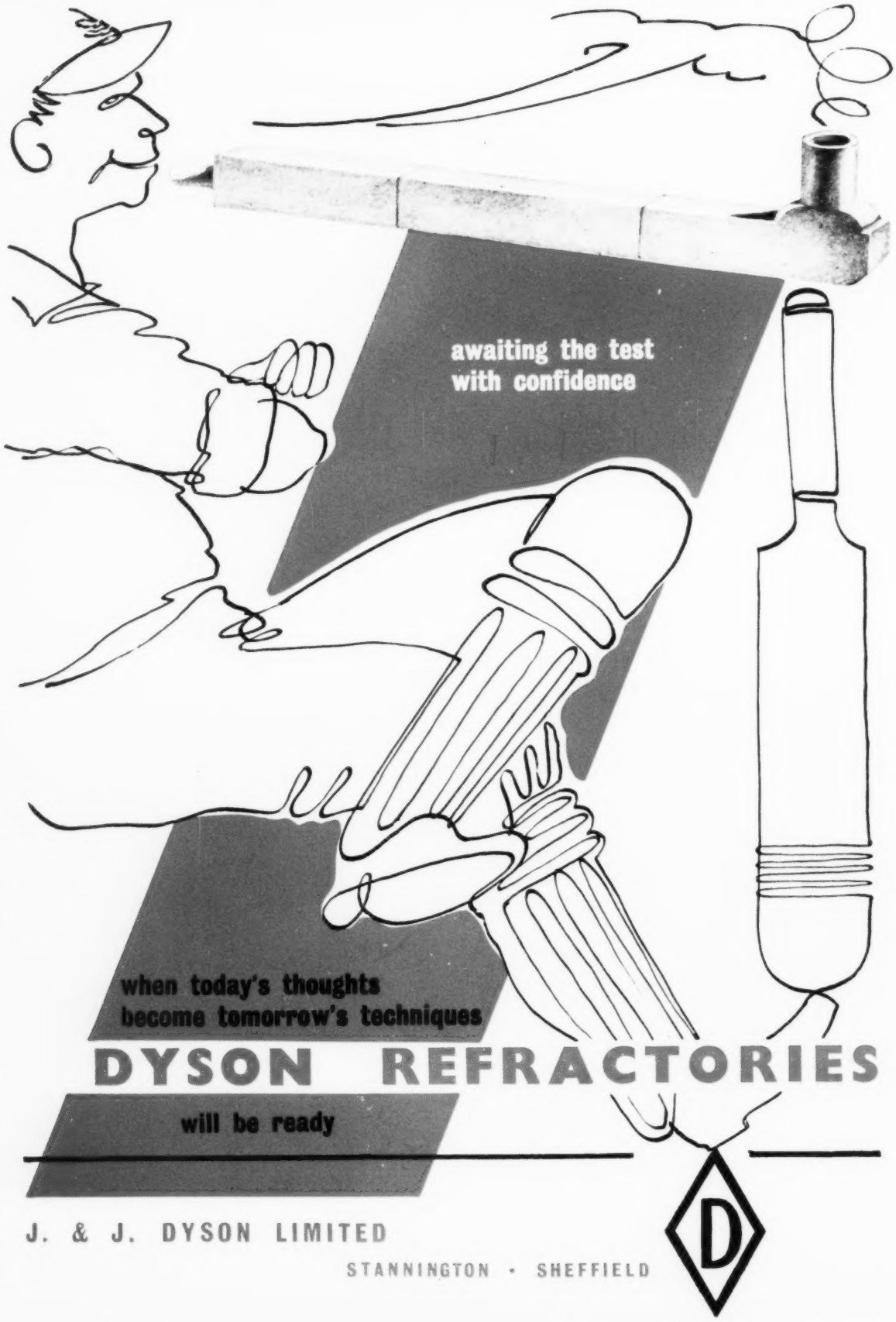
Notch ductility and tensile properties of synthetic mild steels
A. M. Sage and F. E. L. Copley 422

Serrations in stress/strain curve of cold-worked 301 stainless steel at 20 K
J. F. Watson and J. L. Christian 439

Discussion at Autumn Meeting 1959 440

News 475 Abstracts 479

A full list of contents is given on the Title Page (see page 71 of the Advertisement Section)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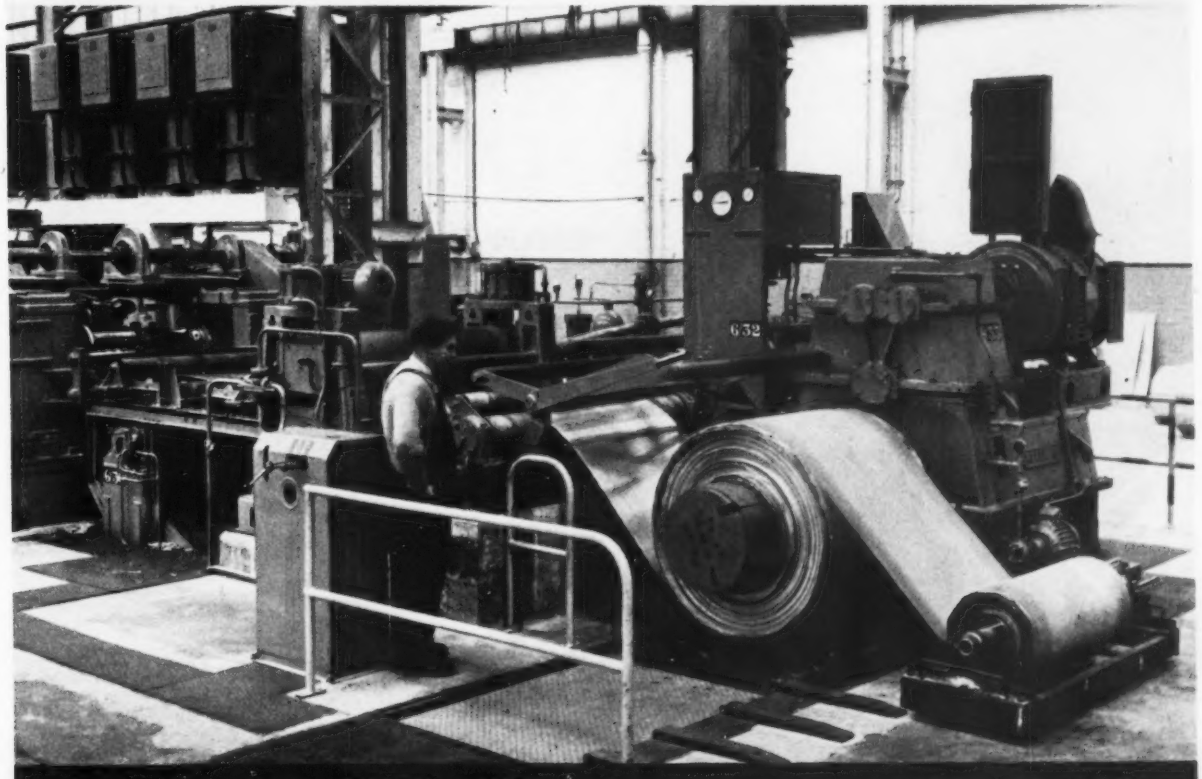

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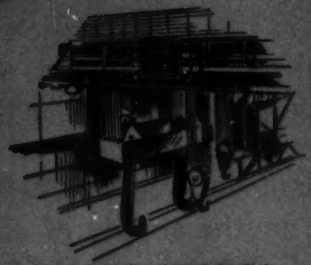

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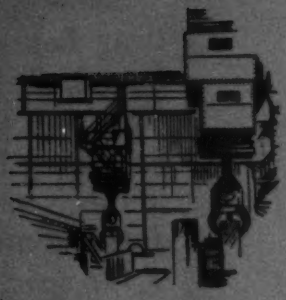

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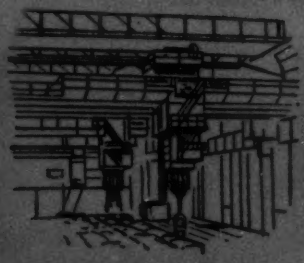

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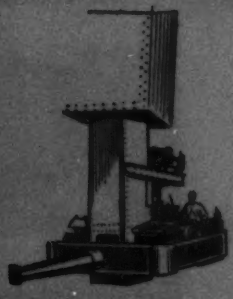

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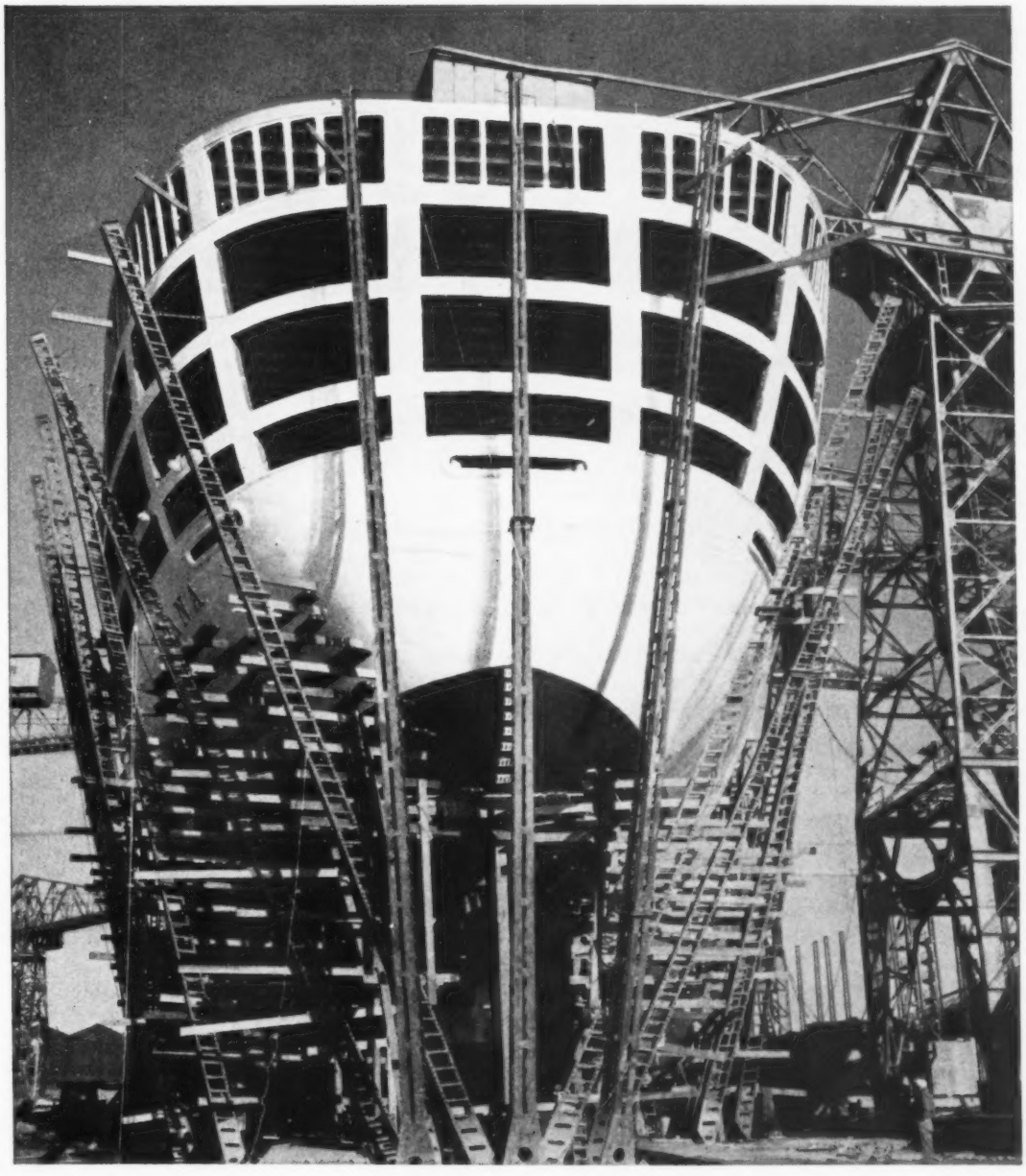

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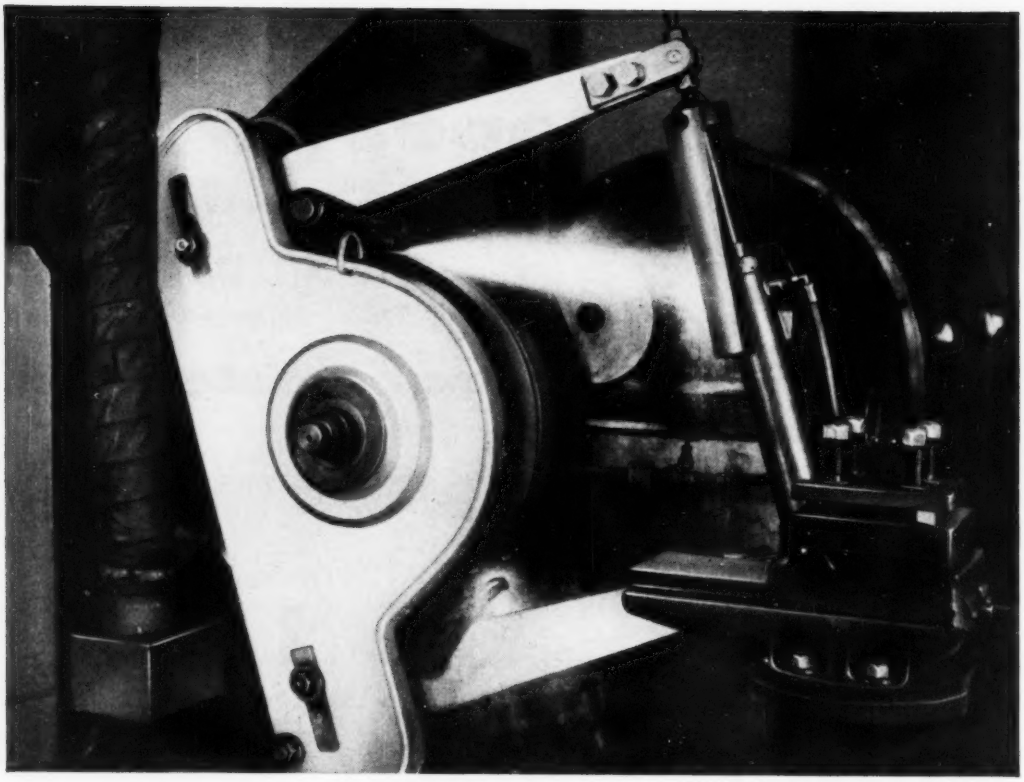
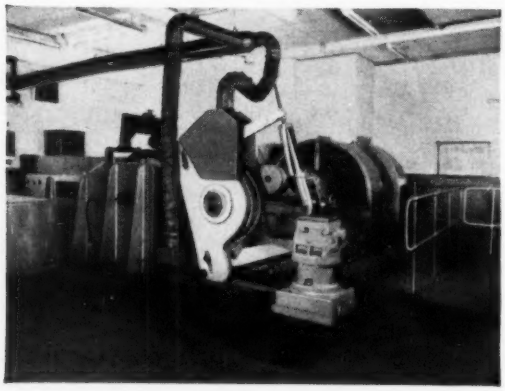


Wheel testing machine braking a solid wheel from high speed.

RAILWAY MATERIALS—
wherever their are railways

Under modern operating conditions railway wheels have to withstand high temperatures, heavy loading and high levels of stress. The wheel testing machine illustrated is used to determine the best qualities of steel for wheels which have to withstand these conditions.



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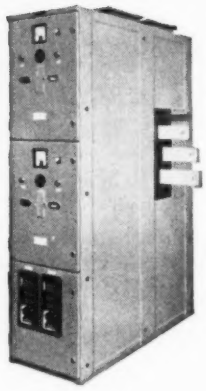

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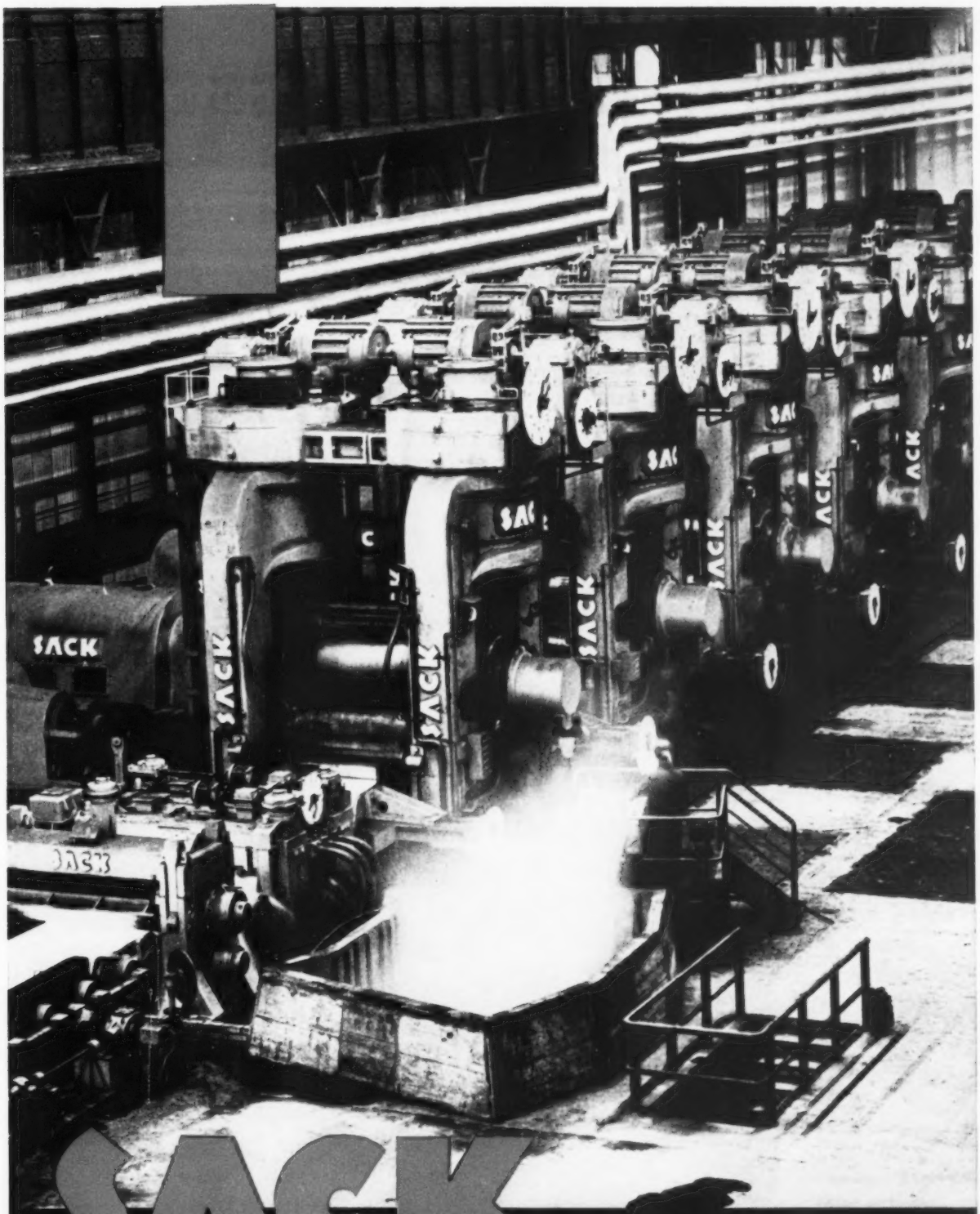

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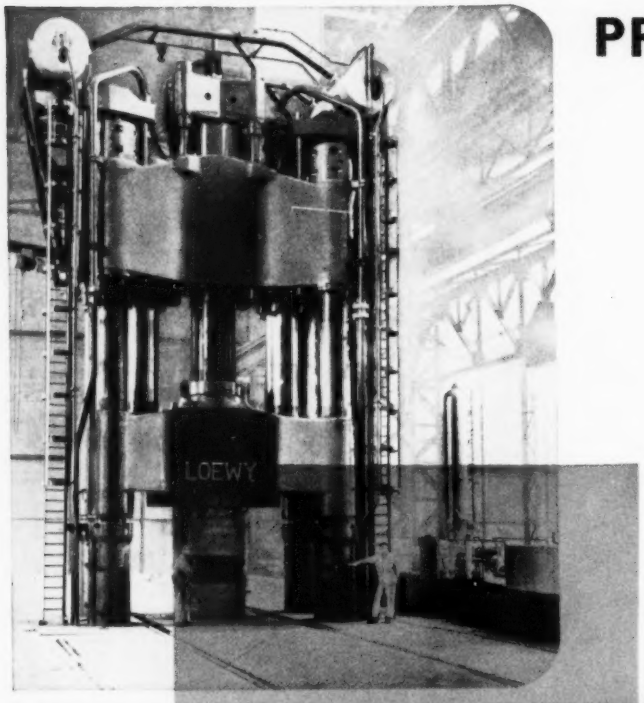

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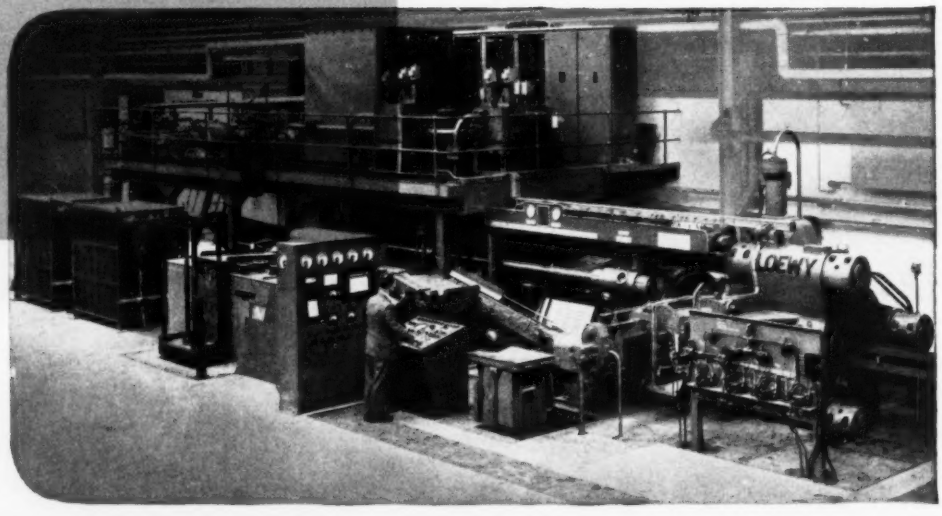

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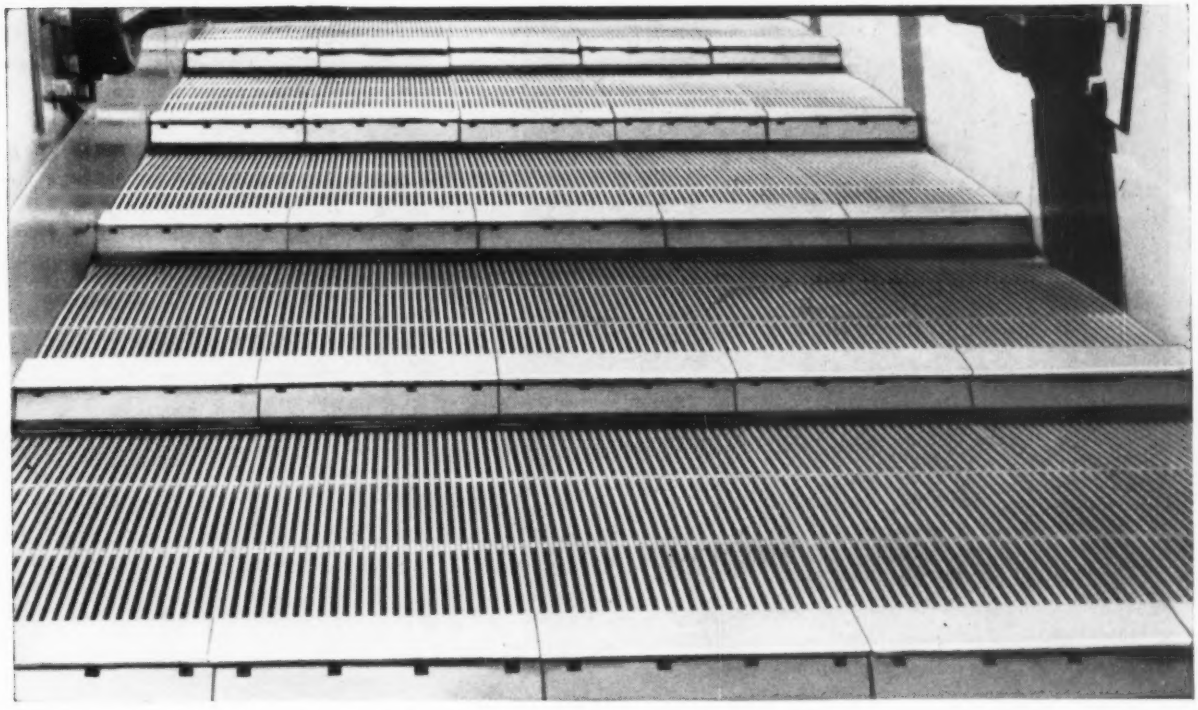

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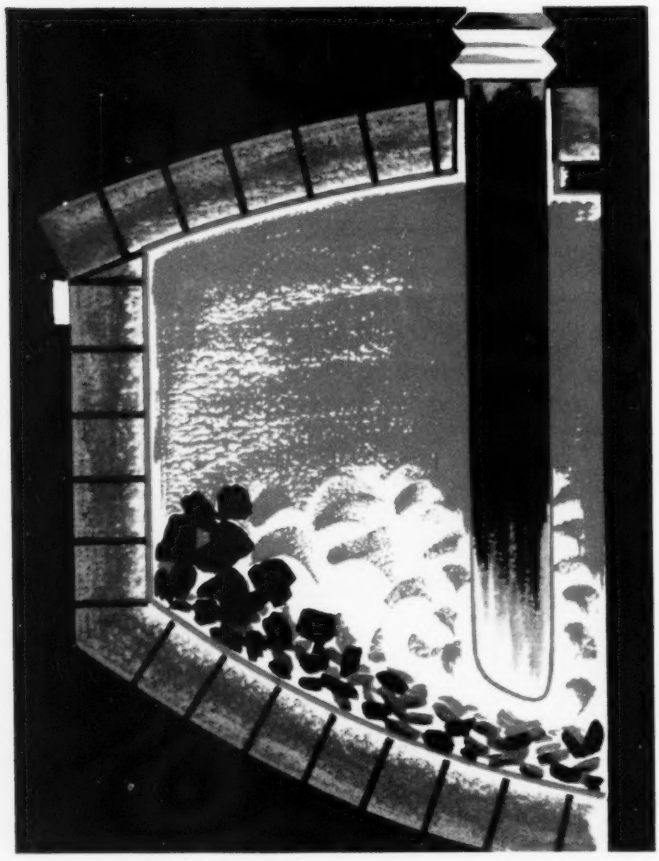

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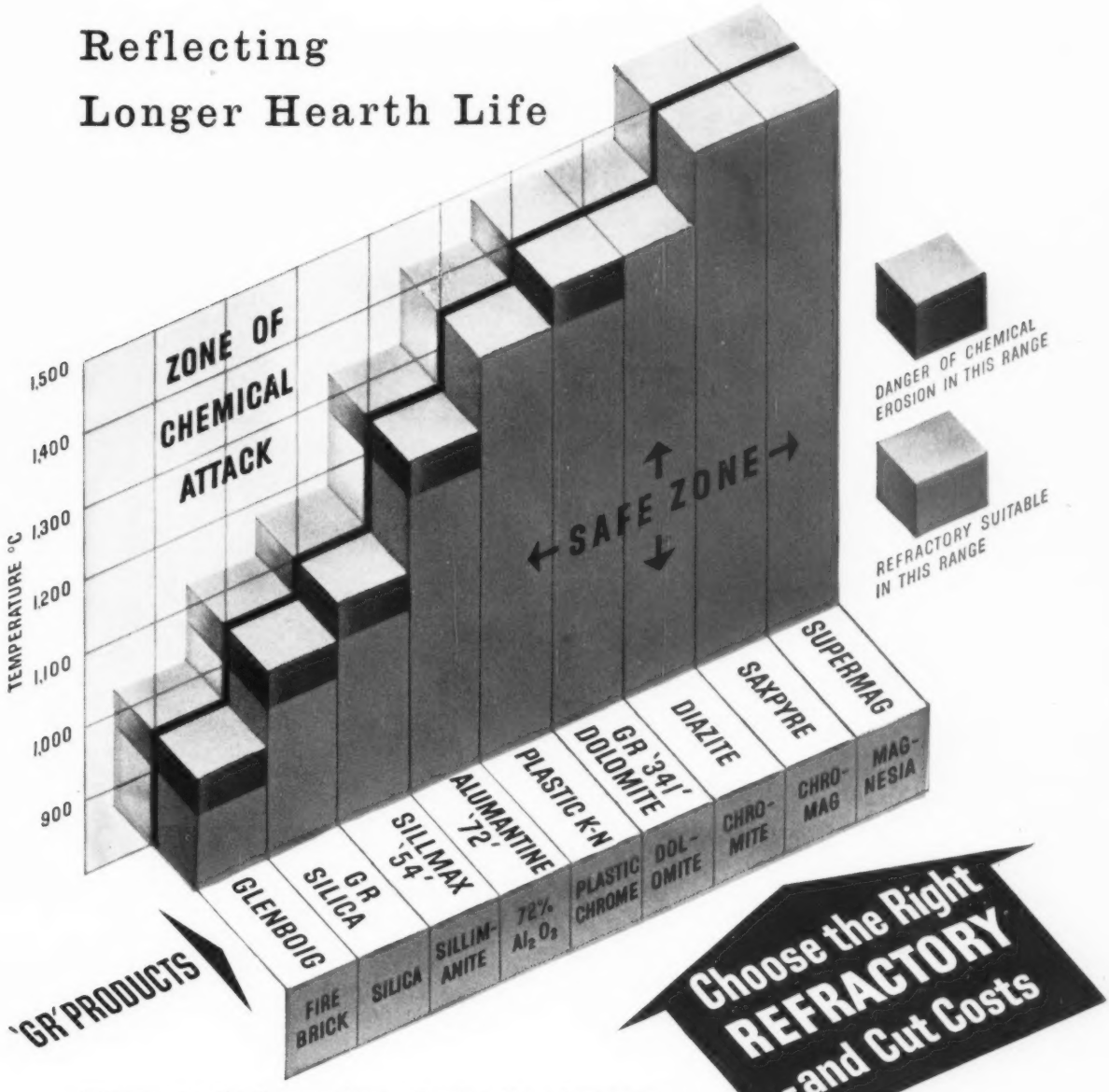

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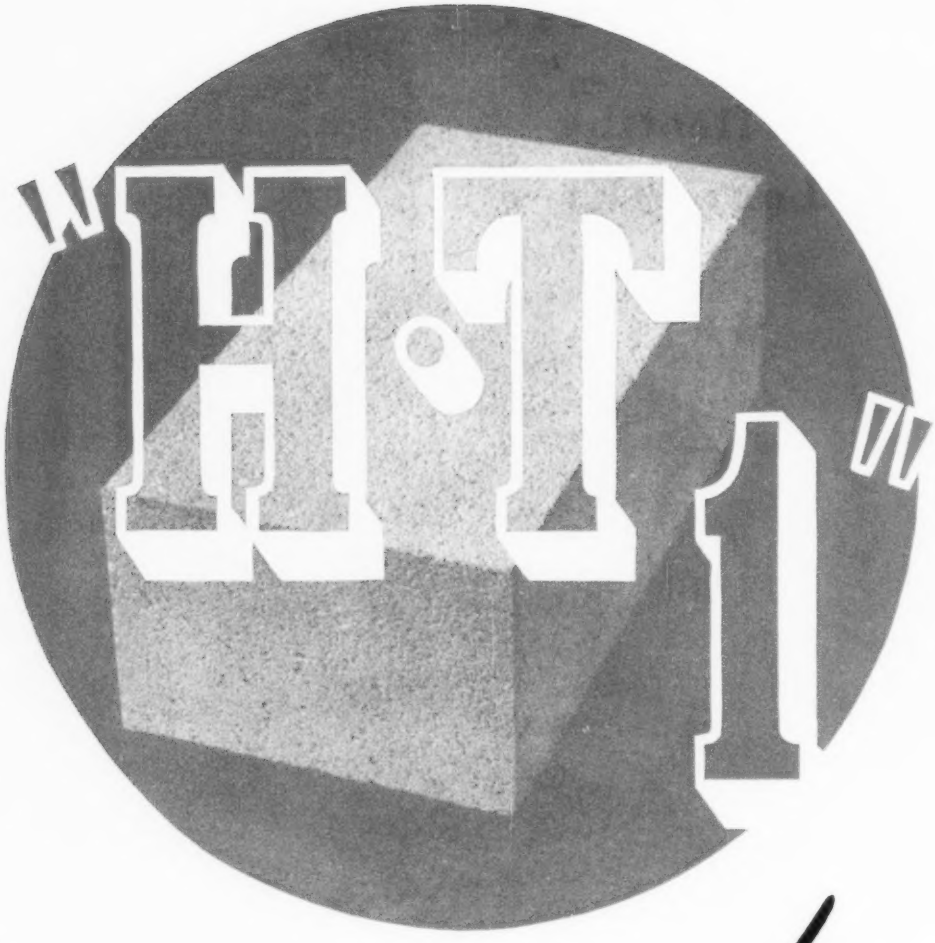

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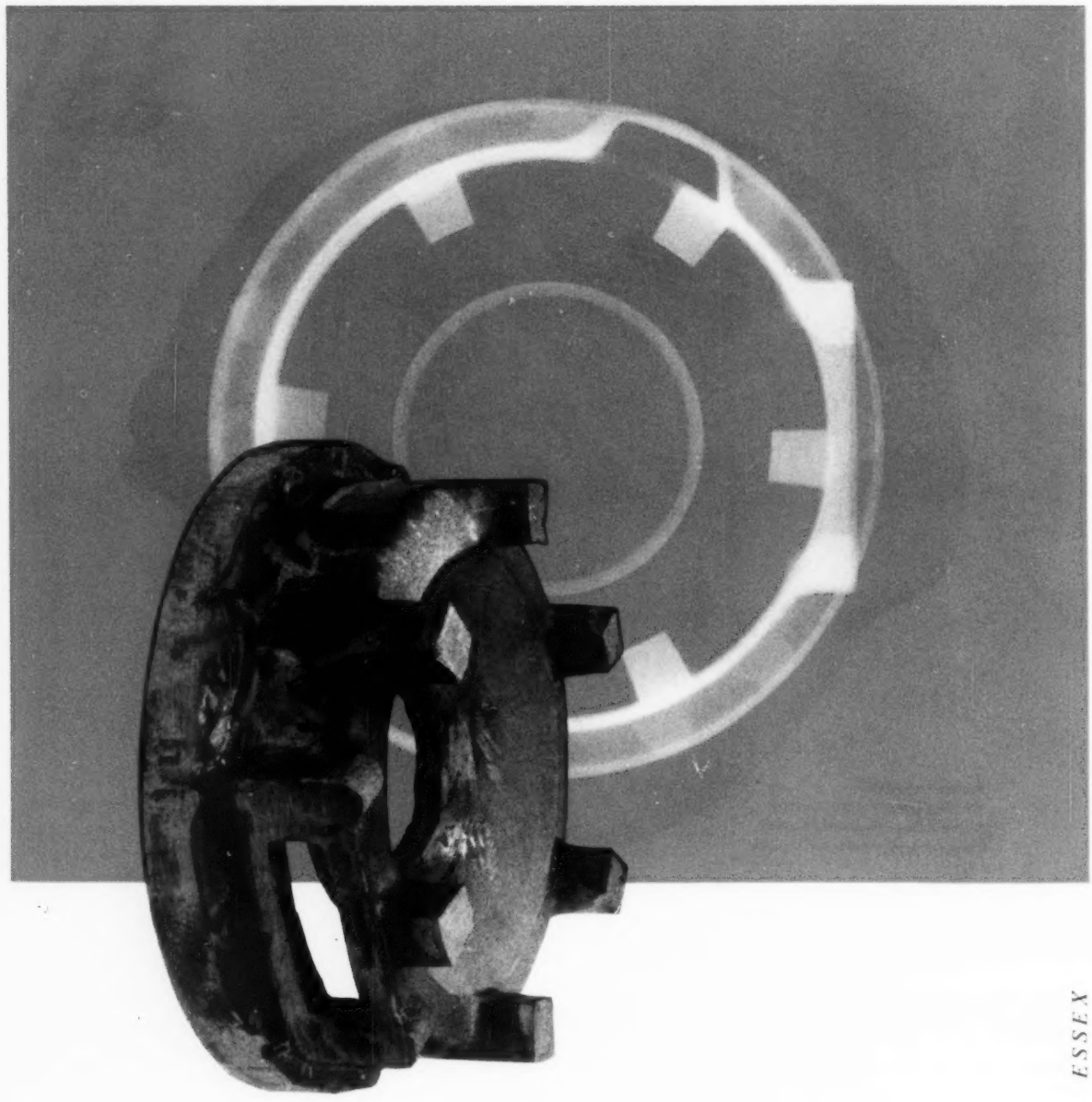

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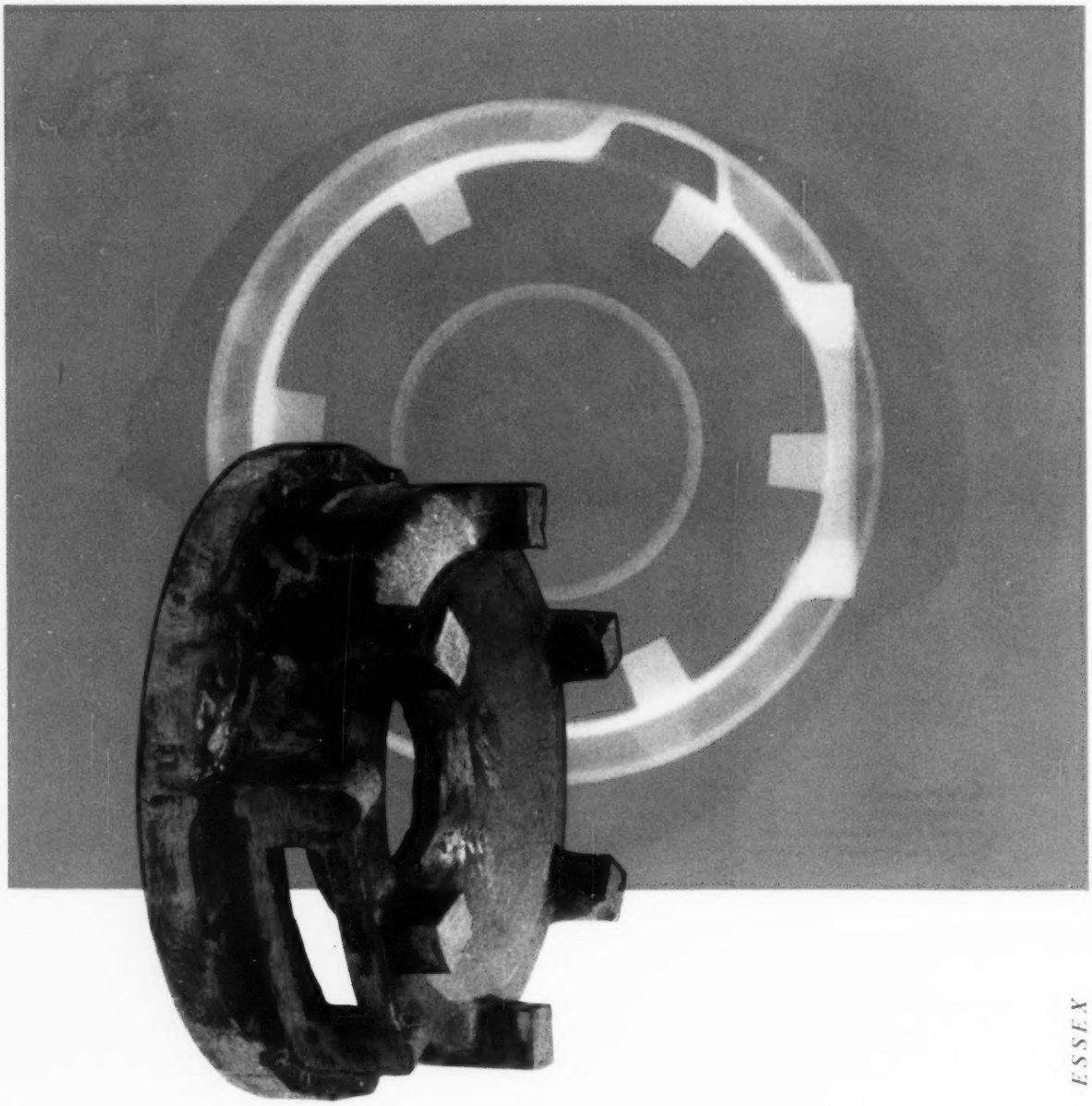

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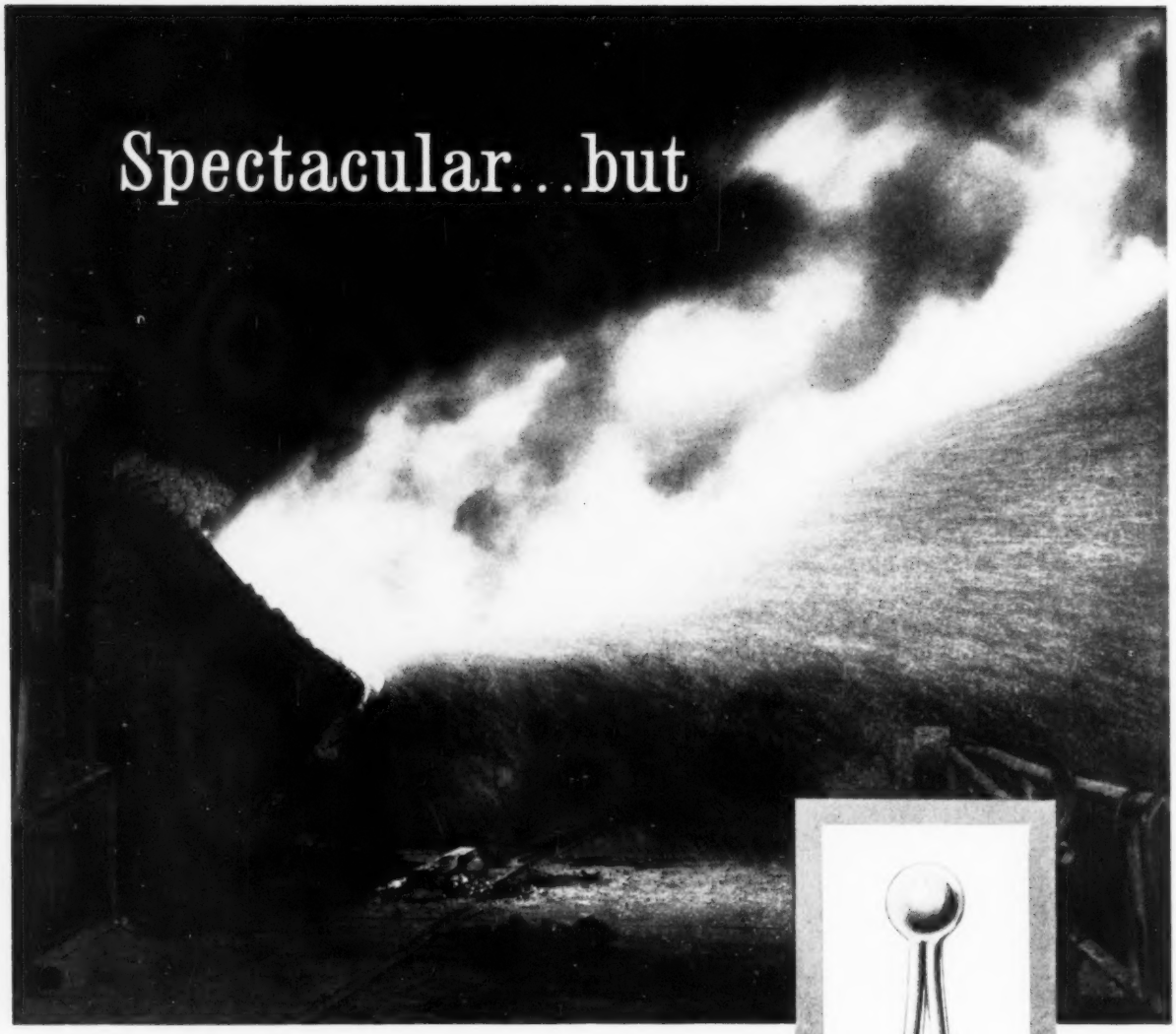

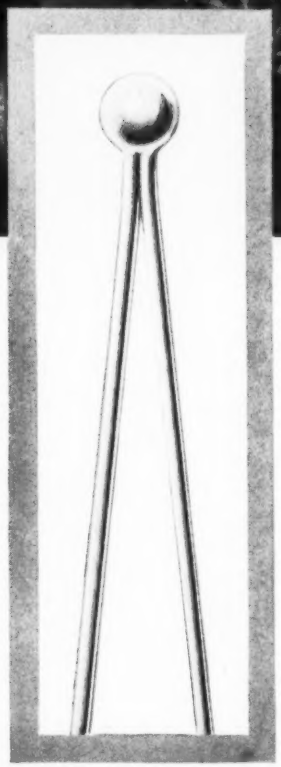

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The First PRECIPITATOR in Great Britain

for LD converter gas cleaning — in fact, the first in the world for the LD/AC method of oxygen application. Installation is in progress at the Ebbw Vale works of Richard Thomas & Baldwins Ltd.

The First PRECIPITATOR in Europe

for oxygen-process, open-hearth furnace gas cleaning, now installed at Scunthorpe for Appleby - Frodingham Steel Co; two further units are on order. The Steel Co. of Wales have also ordered two units.

The First PRECIPITATOR in the World

on a 550 MW pulverised fuel fired boiler unit. It is being installed at Thorpe Marsh Power Station on a boiler built by International Combustion Ltd.

Simon-Carves have produced a new brochure, Steelmaking and Electro-Precipitation, for those with gas cleaning and dust collection problems. To obtain a copy, write to CPP Division (JISI) at the address below.

HIGH-EFFICIENCY ELECTRO-PRECIPITATION BY

Simon-Carves Ltd

STOCKPORT, ENGLAND

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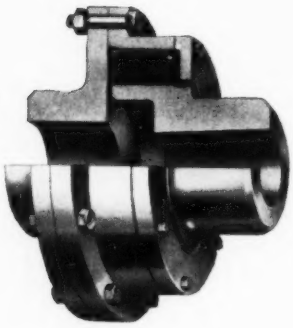
HOLSET FLEXIBLE Rubber Block COUPLINGS

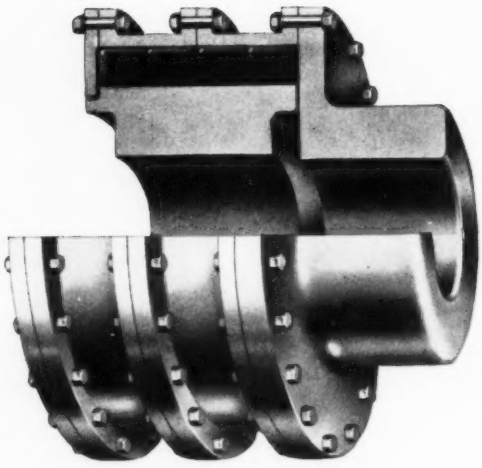
for ROLLING MILL MAIN DRIVES and all AUXILIARIES

CHARACTERISTICS OF HOLSET COUPLINGS

- Can accommodate substantial angular and parallel misalignment.
- Absorb shocks and operate noiselessly under any conditions.
- With a multi-row arrangement of the blocks, Holset Flexible Couplings can have a small diameter and, therefore, a lower WR^2 value than any other torsionally resilient coupling.
- Do not create any end thrust on the bearings of the connected shafts.
- Provide electrical insulation between the driving and driven machines as there is no metal to metal contact.
- Do not require lubrication or maintenance of any kind.
- Installation of Holset Flexible Couplings is simplicity itself. As there are no spigots, removal of the bolts permits vertical lift of the driving and driven machines.
- The design is consistent throughout the range of standard sizes which extends to couplings with a torque capacity of 400 metre tons.

Mill Motor Coupling up to 200 h.p.





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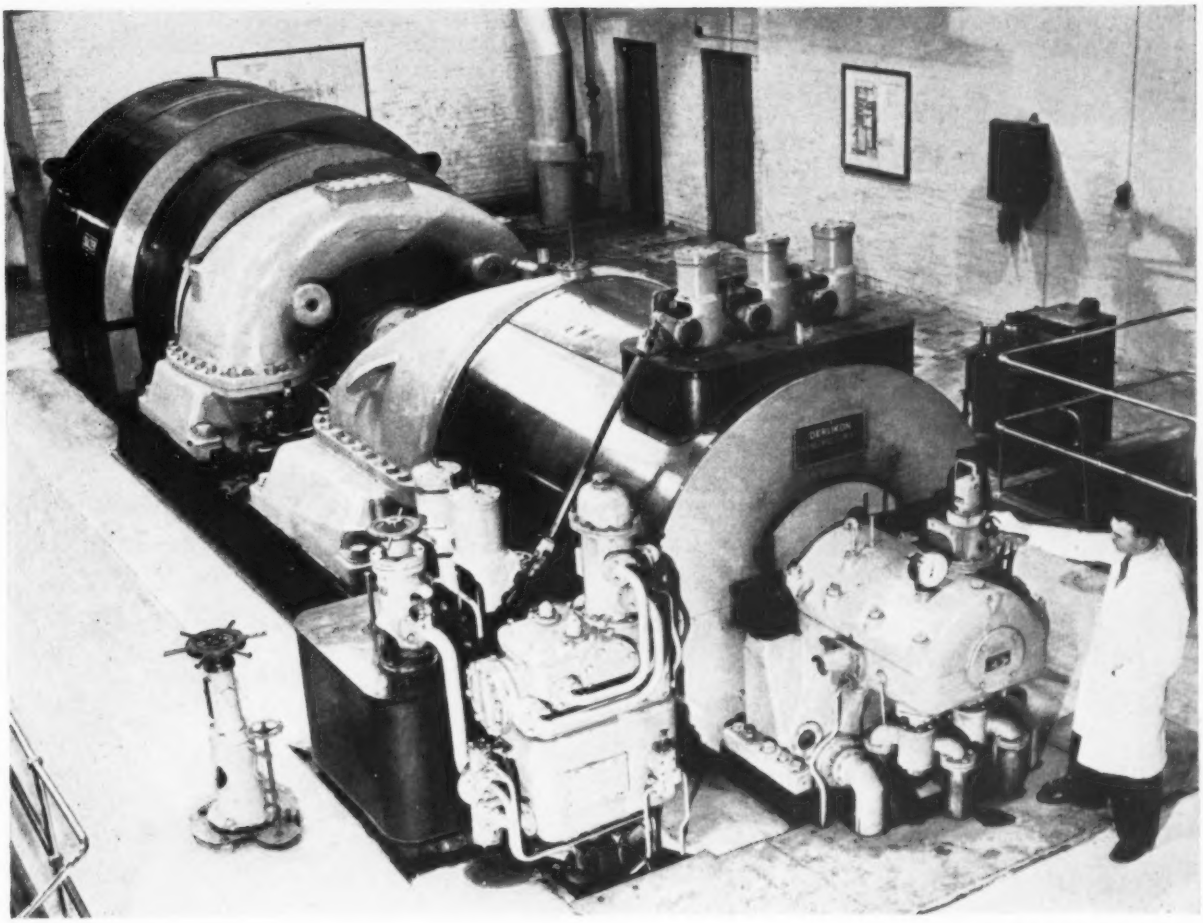

No. 1 Turbo-blower, Margam 'B' Power Station.

Photo courtesy The Steel Co. of Wales

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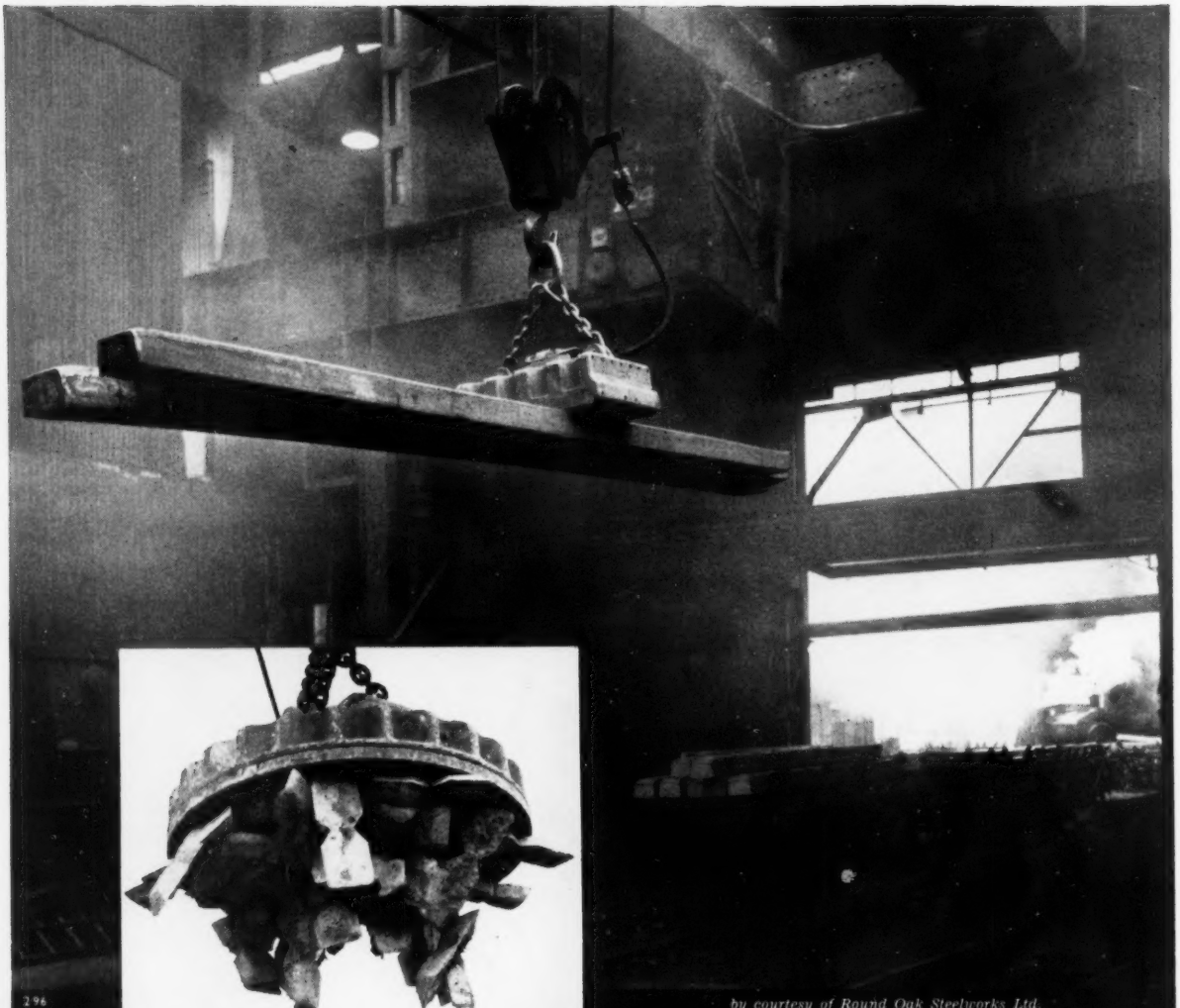

by courtesy of Round Oak Steelworks Ltd.

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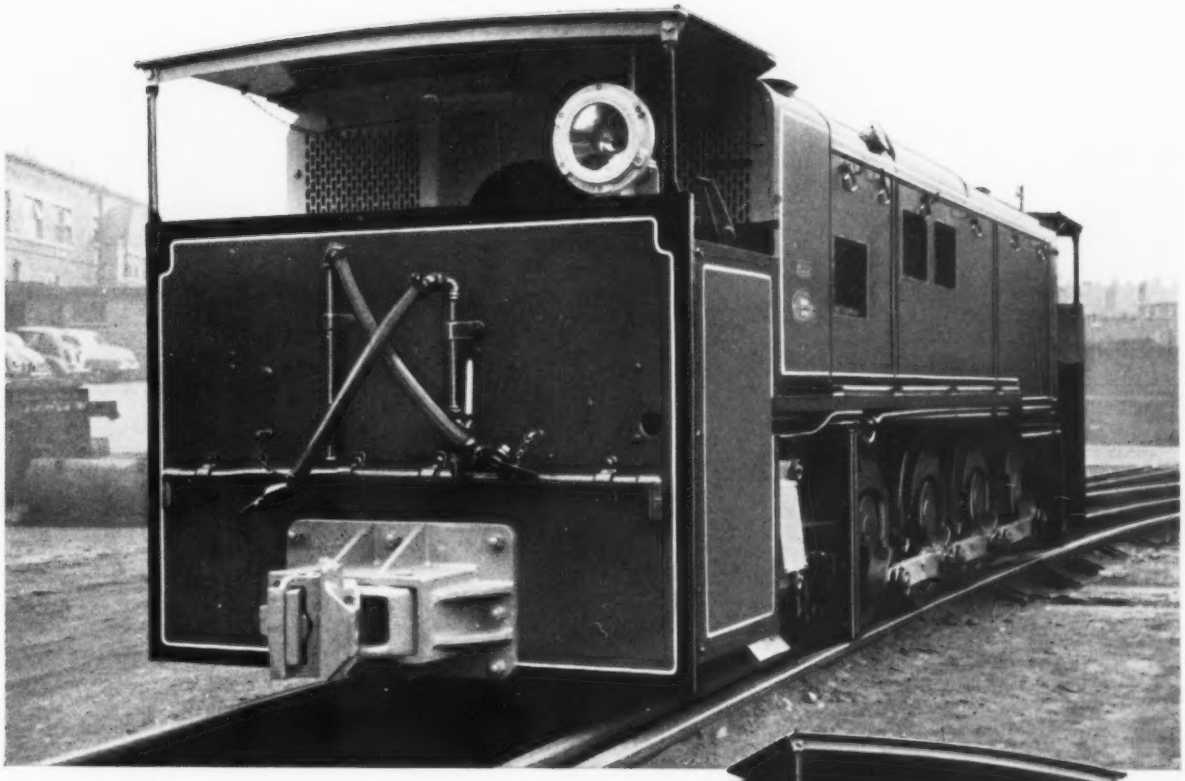

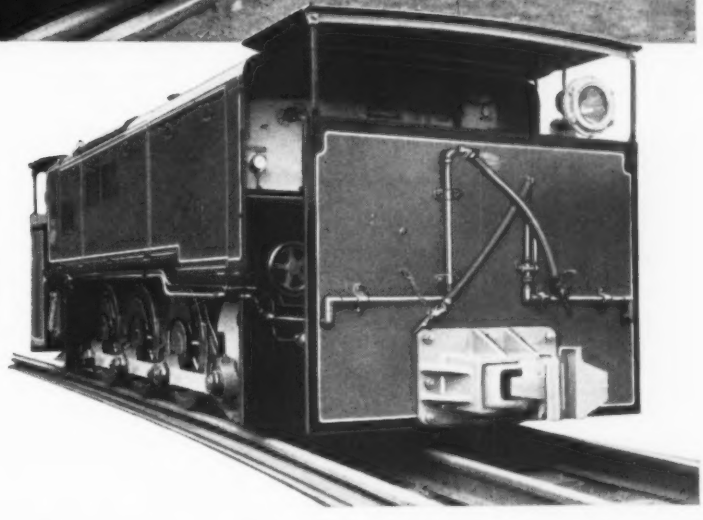

100 H.P., 0-6-0 type, 3′ 6″ gauge diesel locomotive, maximum height 5′ 6″ supplied to Messrs. Stewart & Lloyds Minerals Limited, Thistleton Mine.
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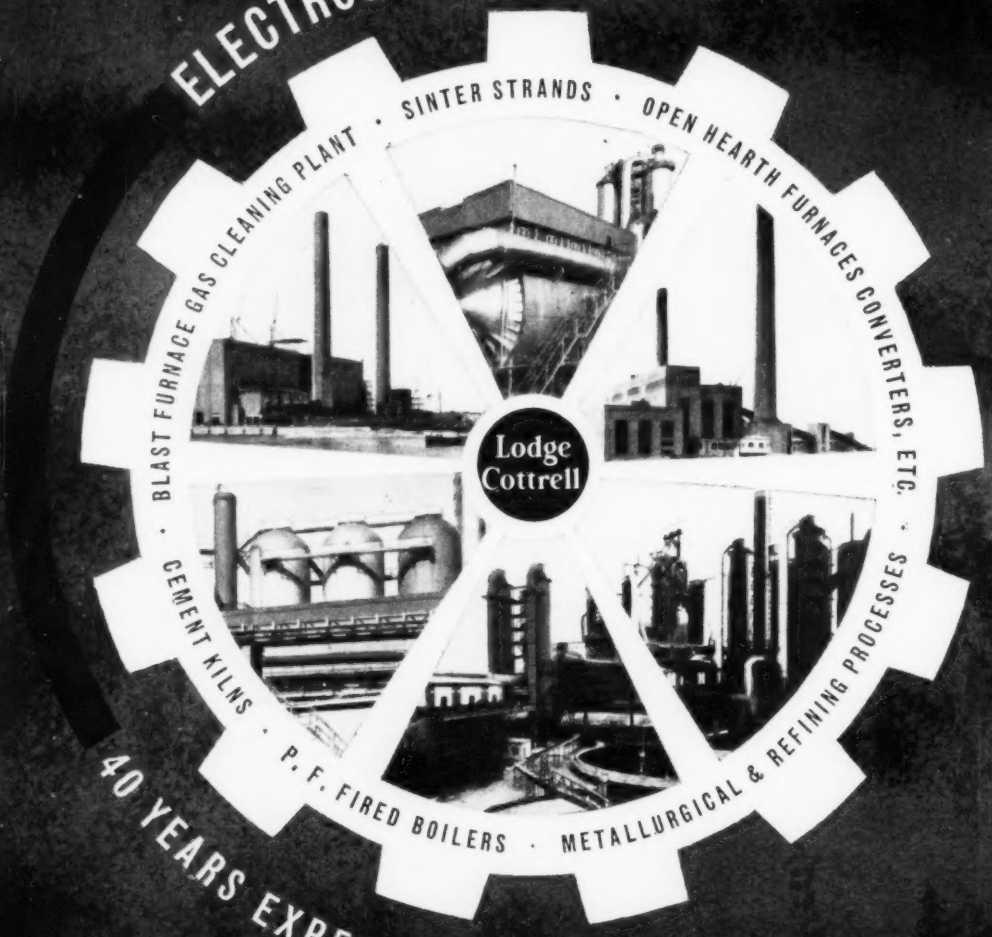
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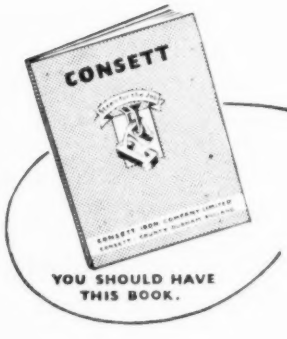

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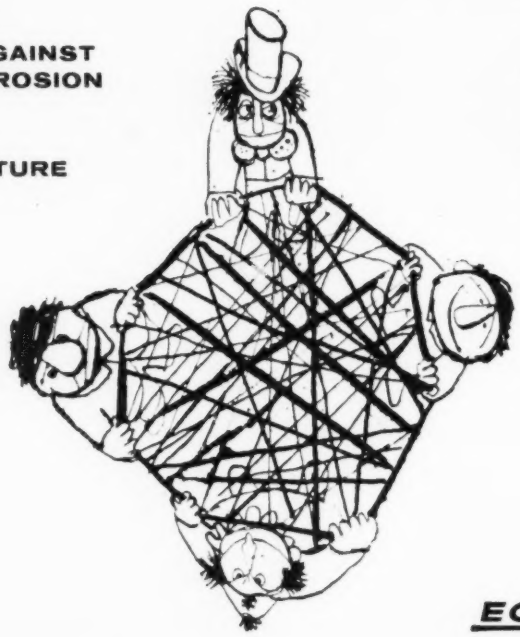

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"Up at AW we like to get on with the job and leave hot air to the furnaces. Nobody has ever managed to *talk* steel into a mould and if our research boys thought that words would make a better Closeloy Roll they would have had us chucking dictionaries into the melt long before now—those lads never miss a trick.

"No! I reckon I'm a lot happier making rolls than I would be making speeches".

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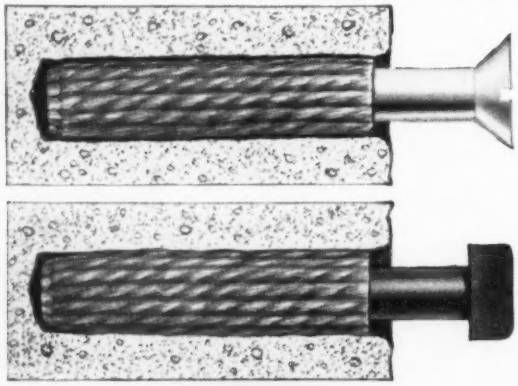

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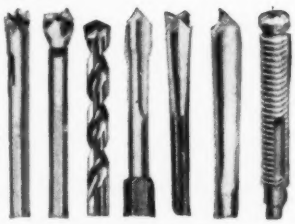

R.P.2 *Two-speed* ELECTRIC DRILL

Keyless Chuck

At the touch of a button you can change the speed of this drill from 1200 r.p.m. to 420 r.p.m. which is the ideal speed for Durium Masonry Drills. Chuck capacity is 5/16". Finger tightening holds the smallest drill in rigid grip which increases with the load yet can be easily released.

RAWLPLUG DRILL HAMMER

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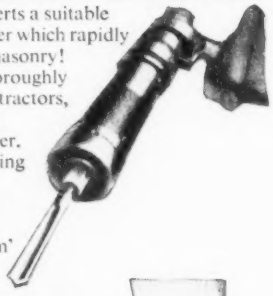

Rawlplug VIBROTO Drilling Machine

The performance of this sensational 3 in 1 power tool must be seen to be believed—three high-precision machines in one! The Vibroto R.V.1 machine provides (1) dual vibration action (with Vibroto Drills, faster masonry drilling even than Durium!); (2) Rotary action for standard Durium Drills; (3) general purpose tool for use with standard twist drills. Vibration frequencies 7,500 and 11,200 per minute. Weight 9 lb. Length overall 18½". Fully guaranteed. The R.V.2 machine is a smaller version and although lighter in weight it has all the advantages of the R.V.1 for drilling holes in masonry up to No. 14 using Vibroto hard-tipped drills.

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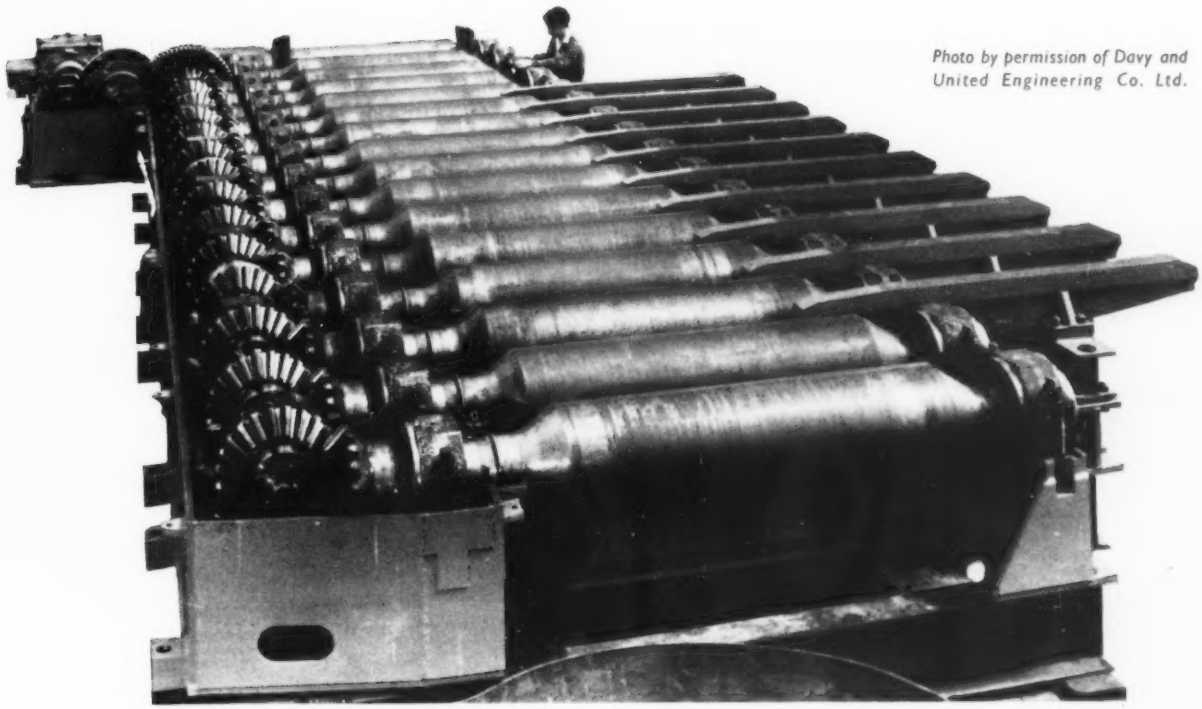

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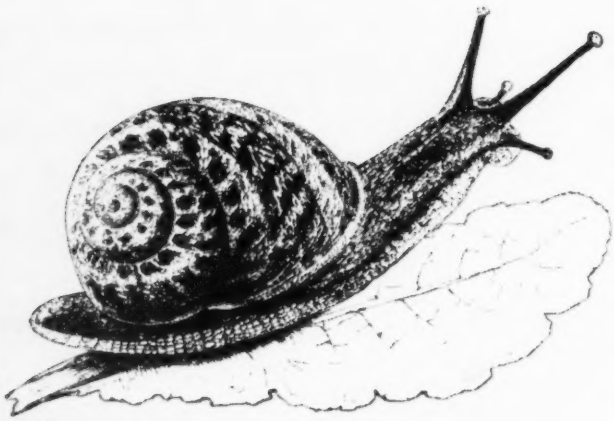

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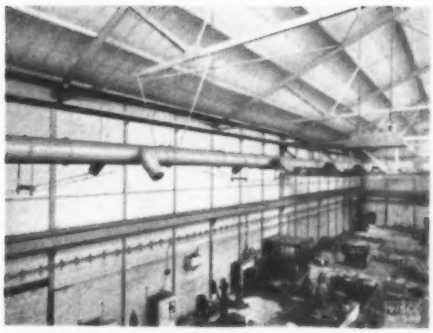
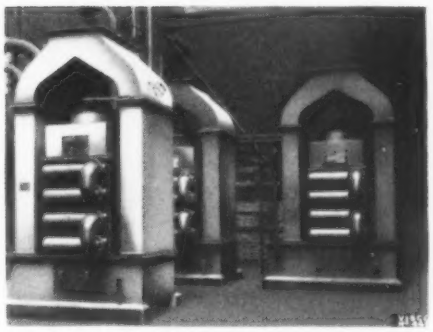
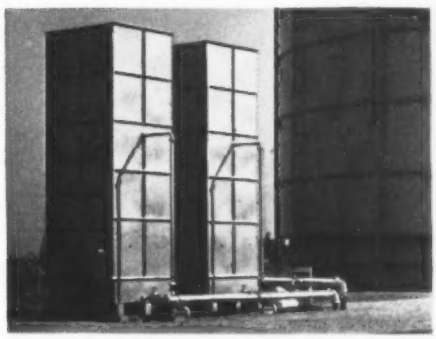


Heating and Ventilating Plant. Firth-Vickers Stainless Steel Ltd.



Visco Spray Blast Oil Coolers. At a large steelworks.



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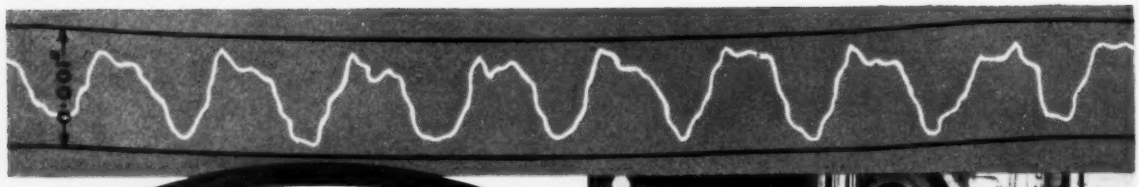

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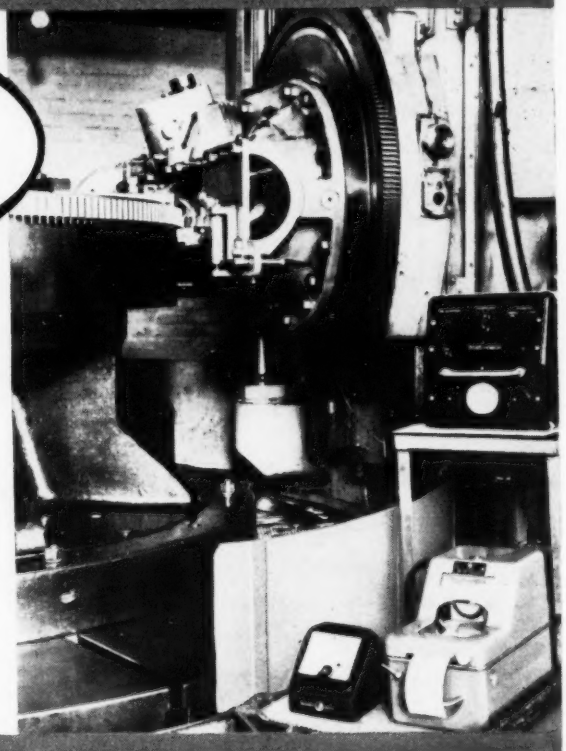

The diagram shows a composite-error record (in part) of a precision-hobbed 4 D.P. spur ring of 56.250-in. P.C.D. Photograph shows gear ring on hobbing machine after cutting, with master gear spring-loaded in close mesh, and error record being taken. *Total error over 225 teeth—including master gear error—0.0012 in.*

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that are proved by performance

It is true to say that whenever absolute reliability of sleeves is required Marshall Sleeves are in greatest demand. The improved "FF" sleeve is the product of continued practical research over the years. It has a high chemical quality and is successfully used in all major steelworks throughout the country.

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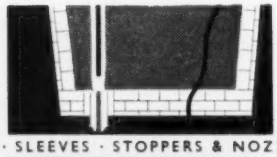

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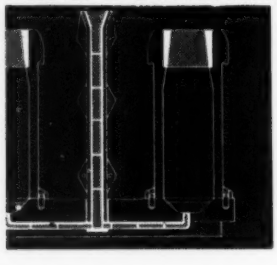

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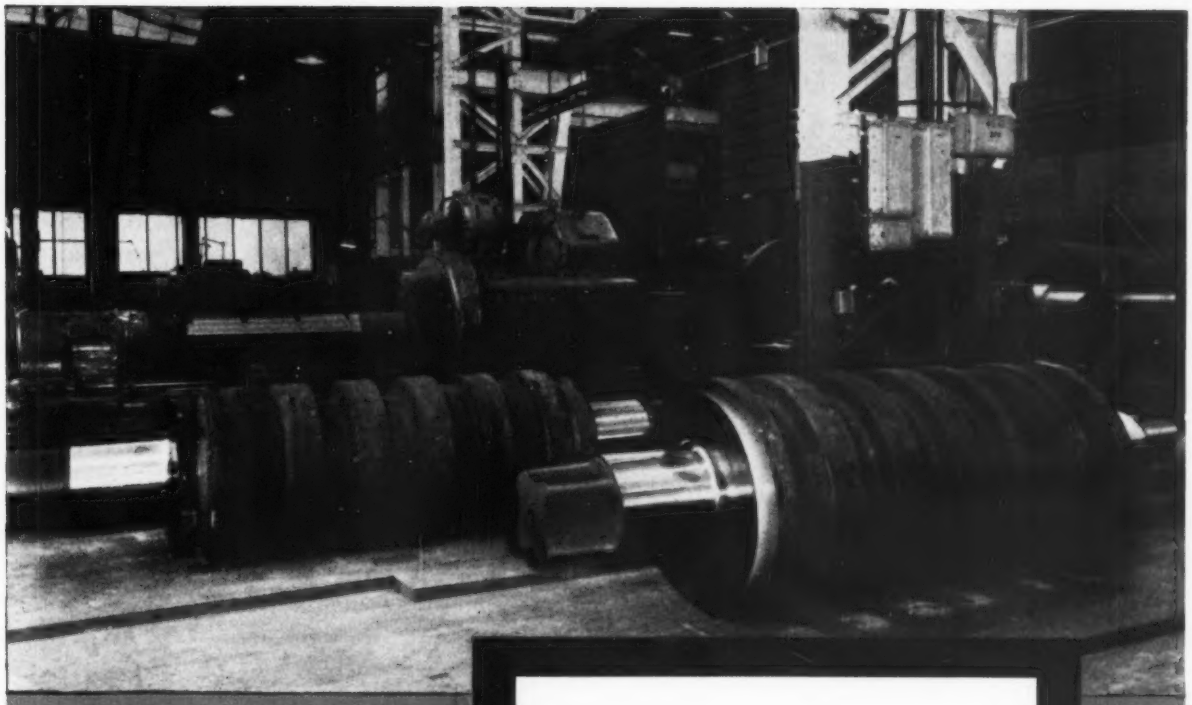

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The Brightside range of rolls for hot and cold rolling is keeping pace with the exacting requirements of the steel and non-ferrous industries at home and abroad.

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present the story of the development of 'Metalkase' bricks

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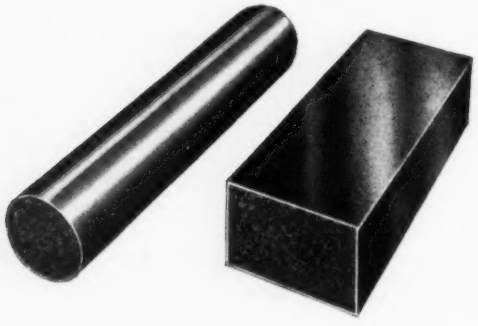
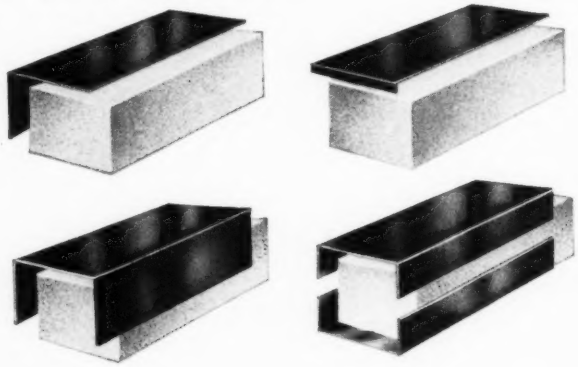
Metalkase tubes manufactured by O.S.F. Co., under British Patent 10598 14, pioneered the development of steelclad basic refractories in the U.K. For over forty years metalkase and metalshape refractories, due to their high rigidity and spalling resistance, have given outstanding performance in open hearth and electric arc furnaces.



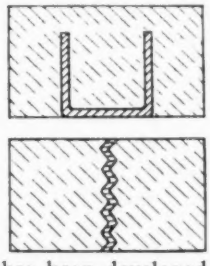
2 Steel Sheet Jointing

The use of steel sheet jointing to produce monolithic basic brick structures has become an accepted constructional method. The welding technique now used by STEETLEY under British Patent 776528, offers to the consumer basic bricks either burned or chemically bonded to which the steel sheet is securely attached as a four-sided case. This speeds installation by eliminating loose sheets and being more reliable removes the dangers during handling which can arise when sheets are attached to bricks by adhesive or as clipped-on cases.

Steel Sheet Jointing (cont'd)

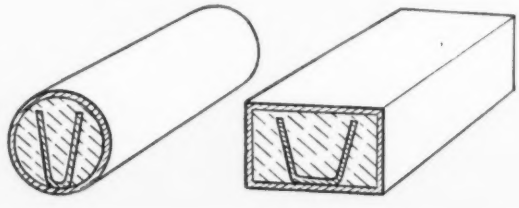


3 Internal Reinforcement



Experience has shown that the performance of metalkase tubes and bricks can be further improved by the use of steel reinforcement embodied in the brick thereby increasing the rigidity still further and improving the spalling resistance by reducing the sectional area. This technique has been developed by STEETLEY and its Associated Companies under British Patents 678637; 690898 and 696311.

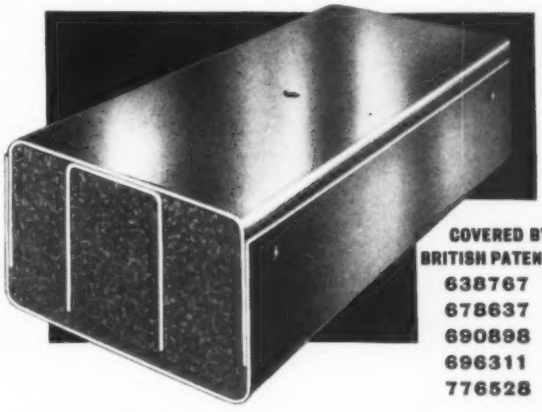
Additional Methods of Reinforcement



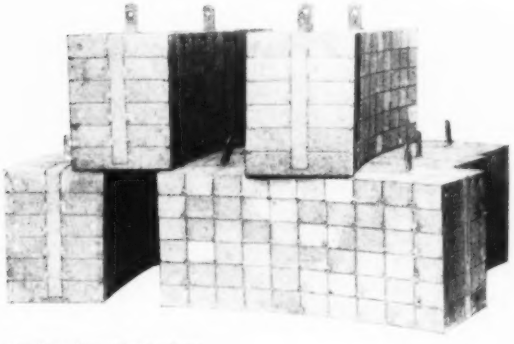
Today... *Steetley present the culmination of their research in 'Metalkase' refractories—*

4 The 'Metalkase' R.W. Brick (REINFORCED WELDED)

As a culmination of this development STEETLEY now offer the R.W. chemically-bonded basic brick combining the principles of welded outer case and internal reinforcement. This brick is available in all the usual compositions and types and whilst being suitable for every basic brick application it is giving outstanding performance in basic open hearth furnace roofs. The R.W. brick can be adapted to any type of roof construction ranging from the restrained arch between fixed skewbacks to the fully suspended flat roof or arch when the bricks are fitted with suspension lugs by extending the reinforcement according to British Patent 638767.



COVERED BY BRITISH PATENTS
638767
678637
690898
696311
776528

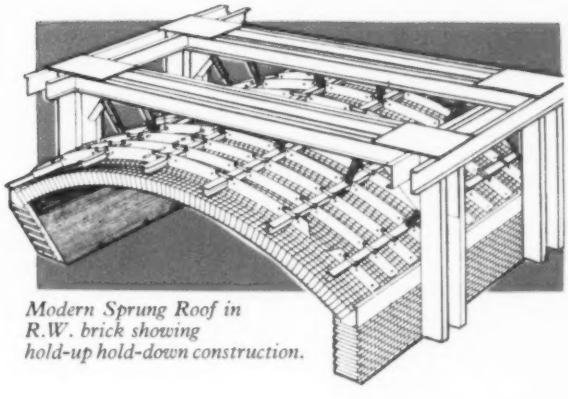


Prefabricated Sections in R.W. brick

For wall construction the R.W. bricks can be used individually preferably as headers or, alternatively, the R.W. bricks can be prefabricated into large sections with still further saving in installation time.

R.W. Bricks offer –

1. **Accuracy of size and shape.**
2. **Speedy installation.**
3. **Exceptional mechanical strength.**
4. **Improved spalling resistance.**
5. **Flexibility of application to all forms of basic roof construction.**



Modern Sprung Roof in R.W. brick showing hold-up hold-down construction.

THE STEETLEY REFRACTORY BRICK DIVISION

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OUGHTIBRIDGE near SHEFFIELD
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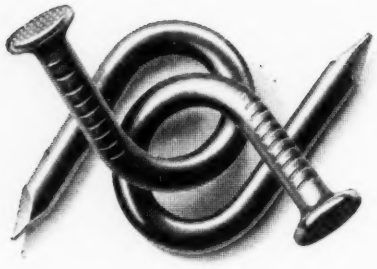
INDEPENDENT INTERDEPENDENCE



The GKN Group believes strongly in what the economists call vertical integration. Expressed more simply, this means doing the job yourself from start to finish. The Lysaght companies provide a good example.

Lysaghts begin with ore. They mine it ; they produce steel ; they operate rolling mills and rod mills ; they have press shops and galvanizing plants ; and they market steel in sheet form and as finished products — principally wire netting, link fencing and farm buildings and equipment. Can integration go farther than this ? It can and does. Lysaghts are also constructional engineers for such projects as power stations, gas works and factories. They design them, they fabricate the steelwork for them and they erect them.

Such relative independence is not unusual in the GKN Group. But, allied to it, there is a strong measure of interdependence between companies. What one cannot do, another can—and what all can do together is to provide a service in steel as complete as you will find anywhere.

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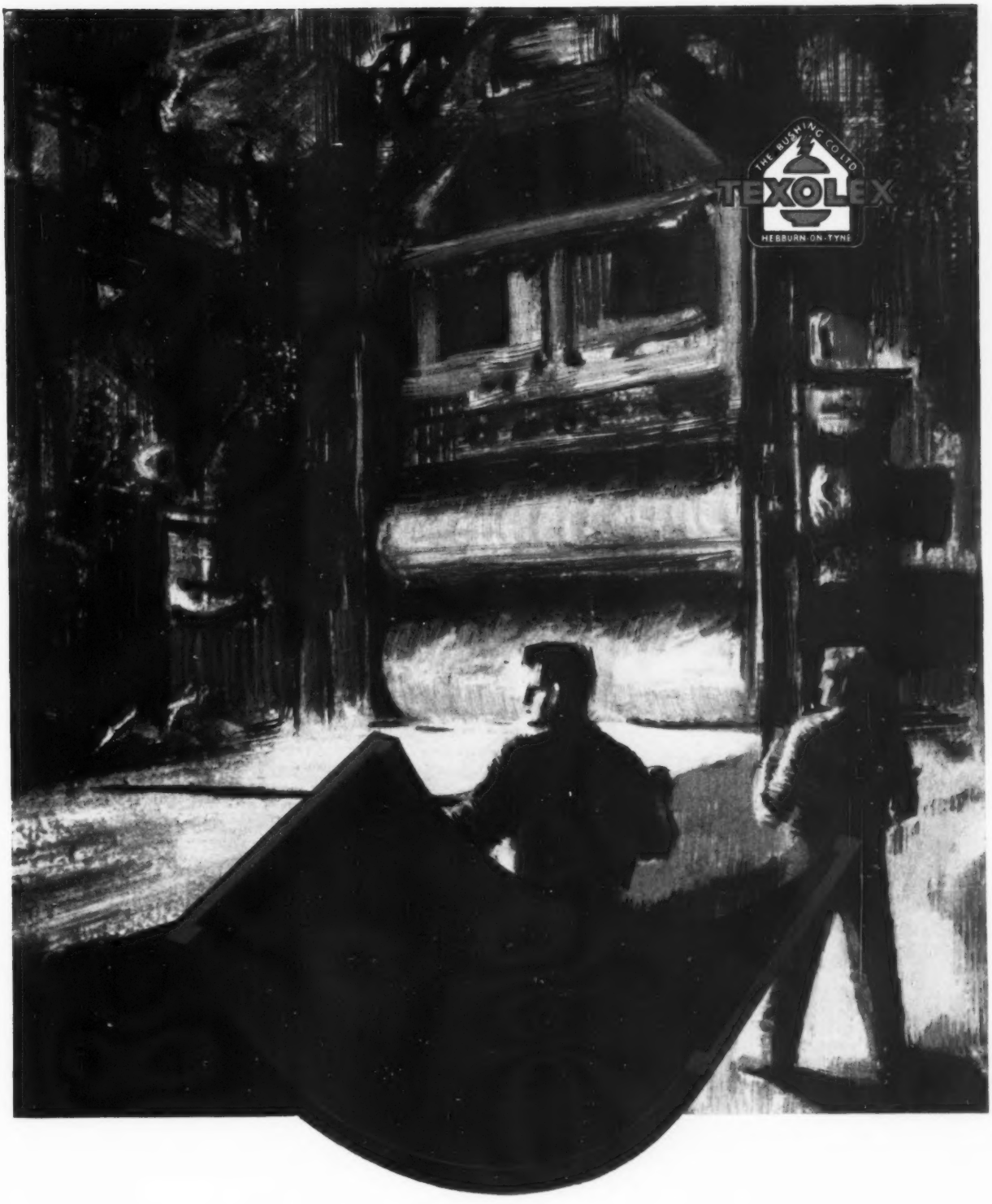

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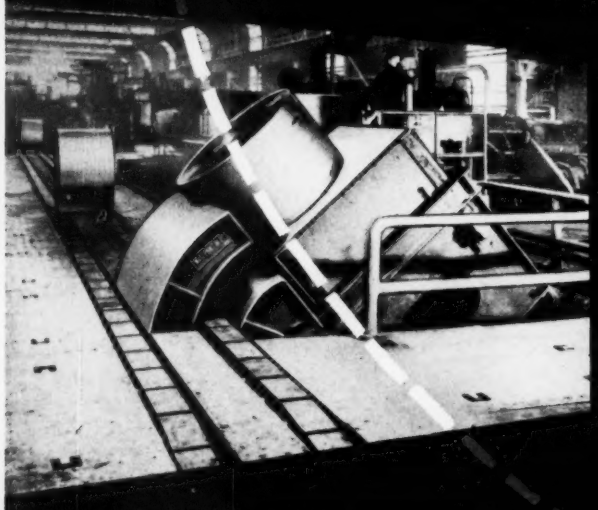
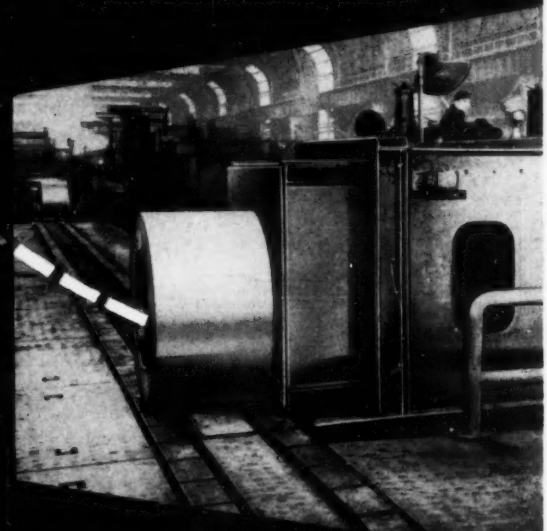
PANTIN

heavy–duty handling systems



Pantin design and manufacture heavy duty mechanical handling plant for the Iron and Steel Industry.

This 'Upender', for example, was made to simplify and expedite the handling of coiled steel. With the utmost ease it takes coils up to 15 tons (max. dimensions 6' diam. × 6' wide) and lays them on a Pantin Pallet Conveyor for movement to the next station.



Pantin Handling Equipment covers every requirement, including upending and down-ending, tilting, hoisting and conveying of coils and sheets.

We are always pleased to discuss, advise and co-operate on the design of handling systems and shall be glad to send you details of Pantin Conveyors and Handling Equipment. Please write for illustrated brochure.

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An 1859 achievement

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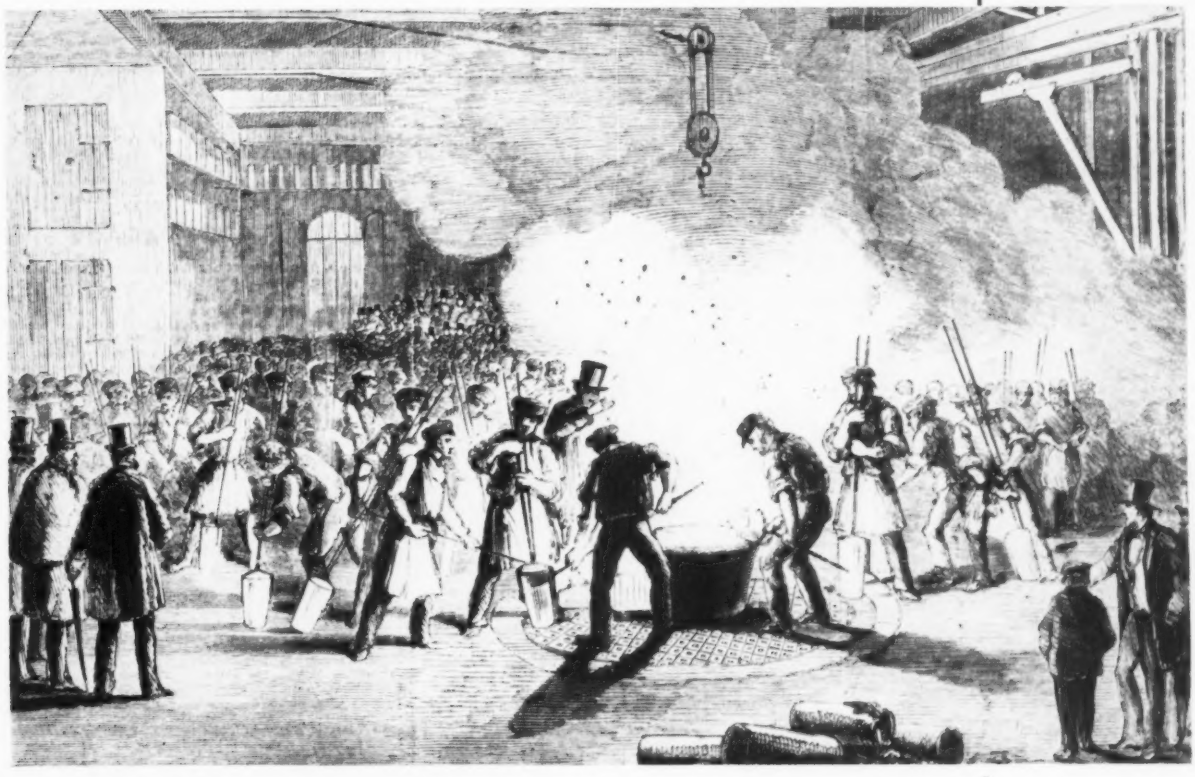
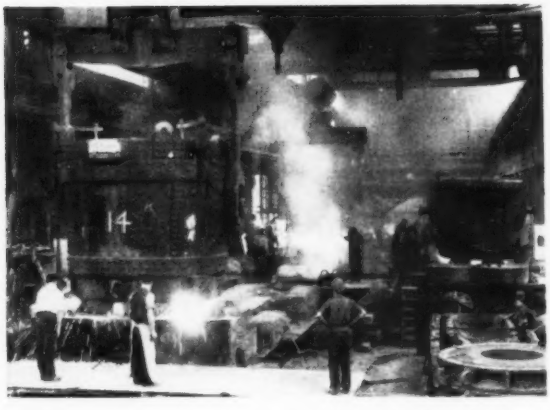


Casting a 75" diameter Bell for San Francisco in 1859.



Pouring 142 ton Steel Casting using three ladles simultaneously in 1959.

and still leading in 1960

With a background of over 100 years in steel foundry practice, English Steel Castings Corporation Ltd., produces a wider variety of steel castings than any other steel foundry in the world (from a few pounds to 185 tons individual weight) and operates the most comprehensive plant in Europe.

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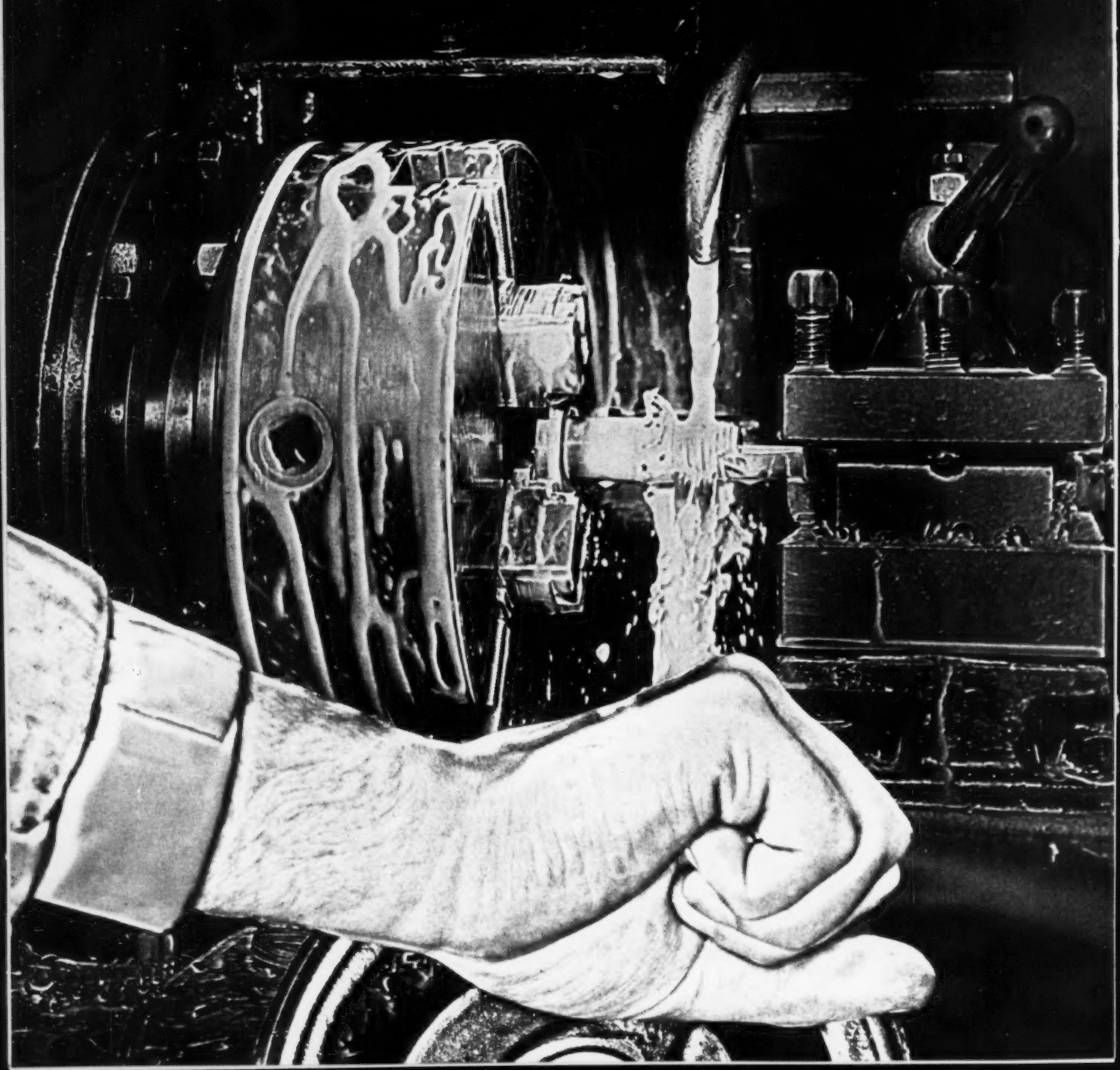


With new Shell Dromus Oils

Most modern soluble cutting oils contain phenolic compounds used as coupling agents between the oil and the emulsifier, for better blending and easier mixing. These phenolic compounds can cause skin irritation, especially where modern high-speed machines are used and the emulsion can concentrate, through the evaporation of water, above the safety level.

Shell research chemists have been working on this problem, which has been causing some concern to Management. After considerable research, Shell Dromus Oils have been reformulated and these new cutting oils now produce bland emulsions, which considerably reduce the risk of skin trouble to operators.

The real difficulty was to find a new coupling agent to replace the phenolic compounds, and Shell finally used what their chemists know as a higher fatty alcohol complex. This solved one problem, but presented another. The new coupling agent was volatile at the high temperatures normally used in blending processes. Further research found a solution to this problem by designing and installing new plant.

The new Dromus Oils are every bit as efficient as before and cost no more. They put Management in the welcome position of being able to minimise working hazards at no extra cost. And machine men need no longer be so worried about skin troubles.

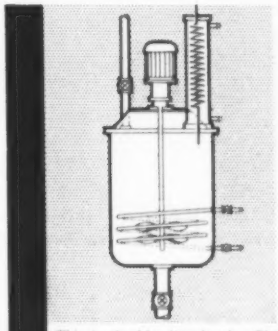
The moral of the story is that Shell research is supremely applicational. The centre at Thornton is always ready to work with even the most specialised sectors of industry to produce the right oil for the job. If you and your organisation have any major lubricating problems, it pays to get in touch with your local supplier of Shell Industrial Lubricants.

The Research Story

Shell chemists in the U.K., in Holland and in the U.S.A., prepared and examined hundreds of experimental soluble oils, and established that certain combinations of fatty alcohols could be used in place of phenolic compounds with no loss of efficiency. They set to work to discover the best combination and developed a higher fatty alcohol complex which fitted exactly. Then they realised that to blend this new coupling agent into soluble oils would require special plant and new blending techniques.

Exhaustive testing of blend stability, emulsion stability, anti-corrosion and machining properties led to selection of the most promising blends. A pilot plant was set up to produce batches of these for use in field trials.

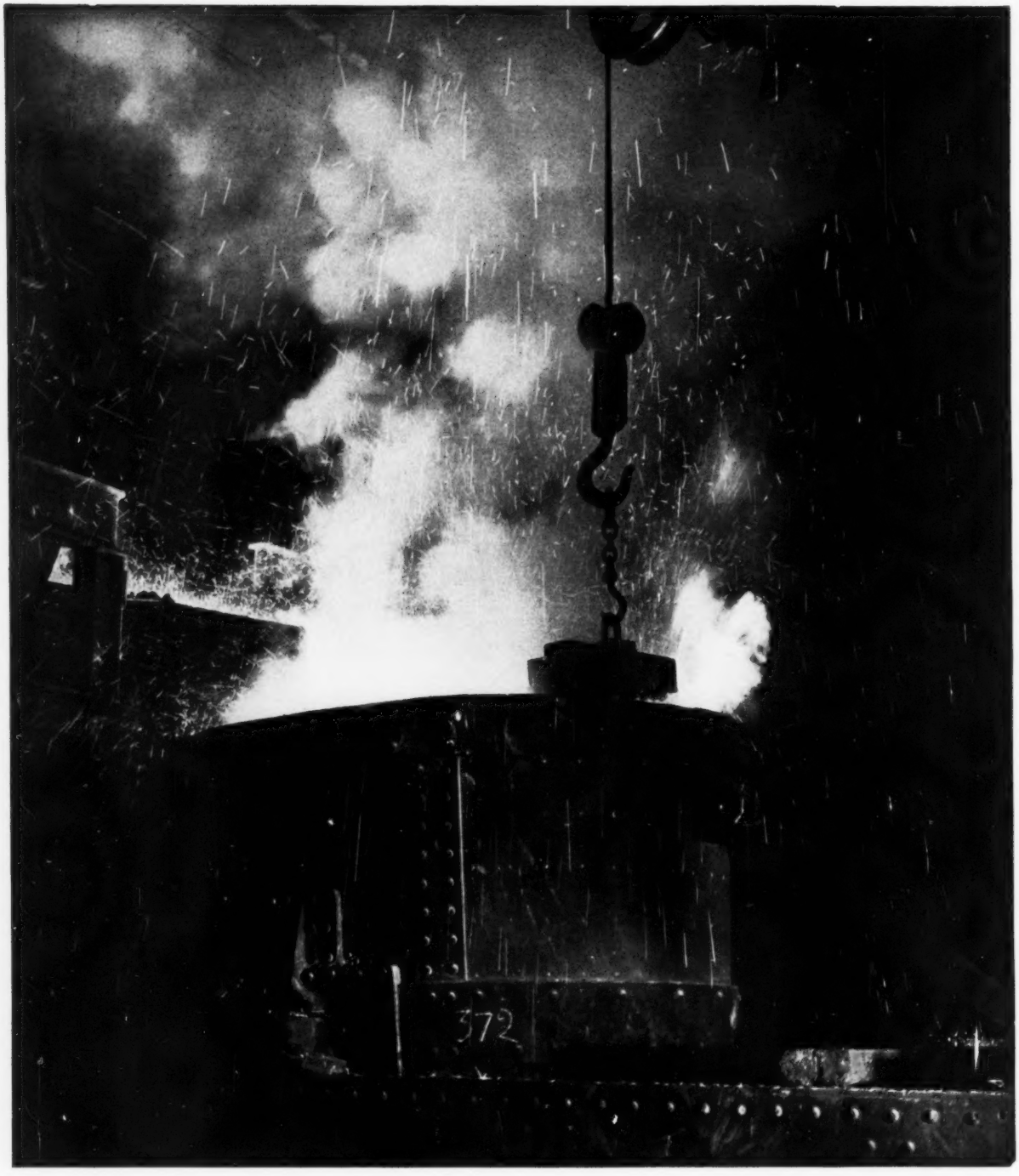
This field testing and final development proceeded for two years whilst production plants were erected at points so chosen as to give the most economical and rapid delivery throughout the United Kingdom.



This is the blending kettle. The reflux condenser beside the stirrer motor prevents the loss of constituents volatile at the blending temperature.

NEW SHELL DROMUS SOLUBLE CUTTING OILS

another proof of Shell leadership in lubrication



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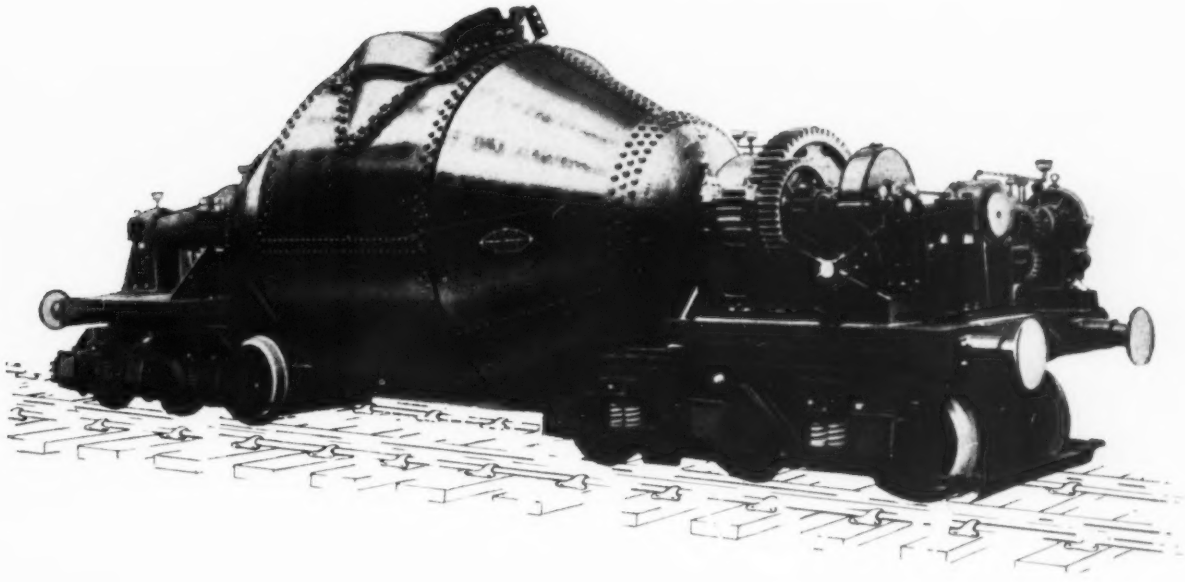

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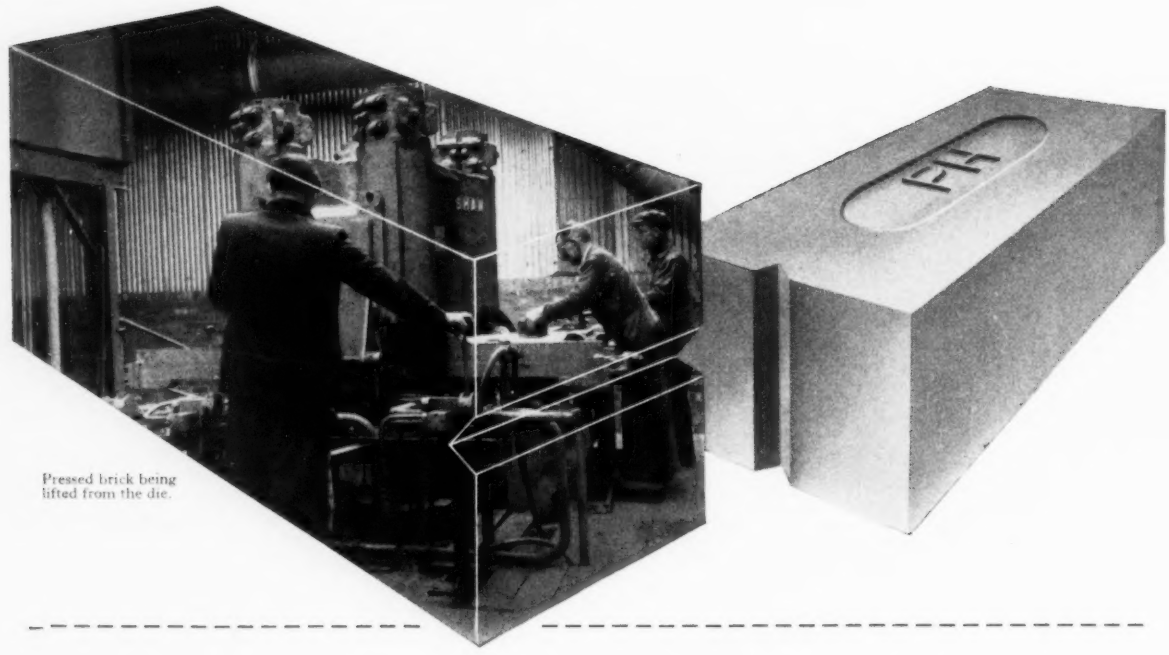

Pressed brick being lifted from the die.

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Refractory bricks today must be absolutely accurate in size and shape. These qualities, combined with unerring consistency in texture and performance are achieved by Pickford Holland through the medium of the most modern plant and equipment.

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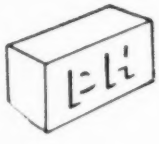
More and more Pickford Holland refractory bricks are being supplied to steel and other industries throughout the world and the demand still grows. This surely points to the success of this policy of plant modernisation and is a tribute to the lasting service that these bricks give.



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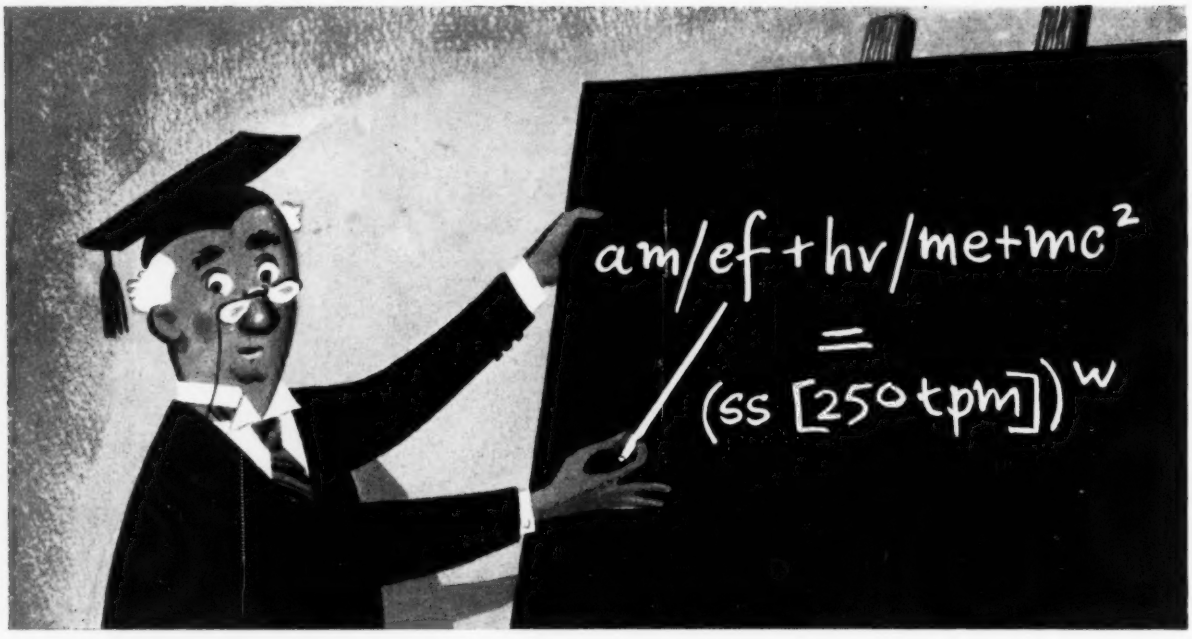

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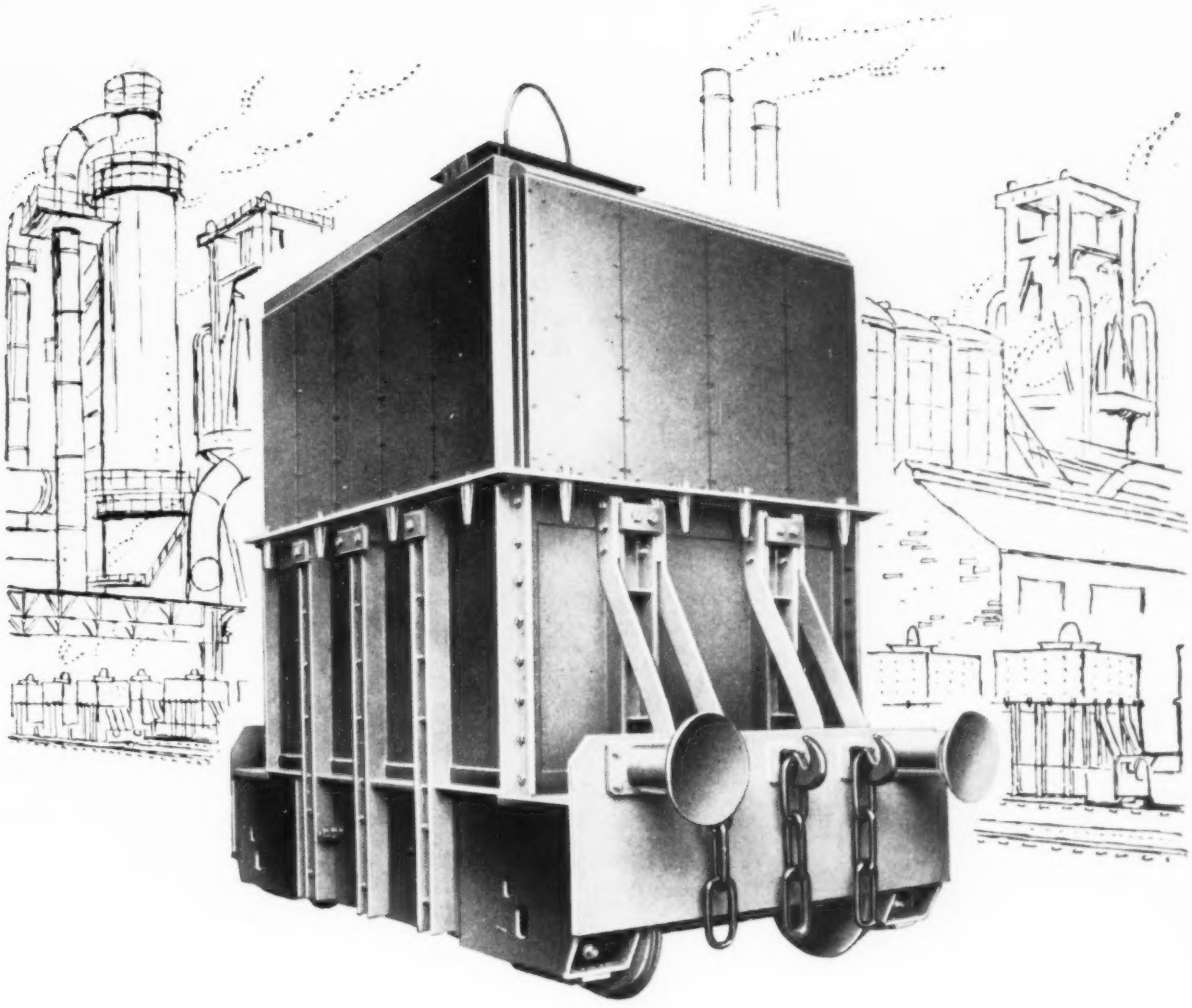

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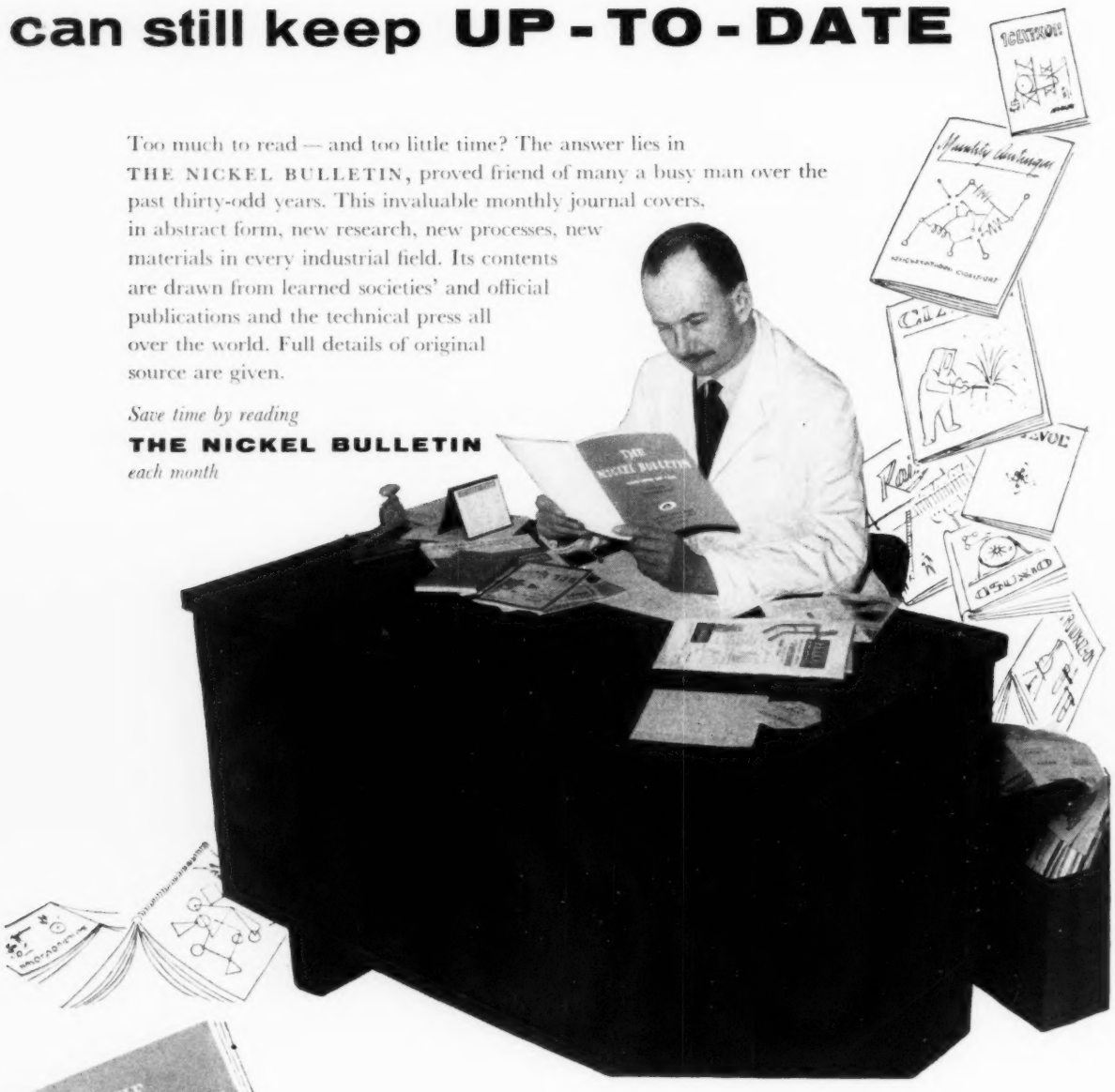

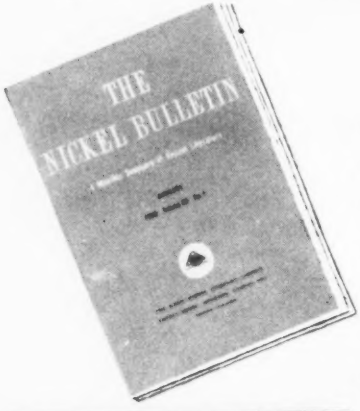

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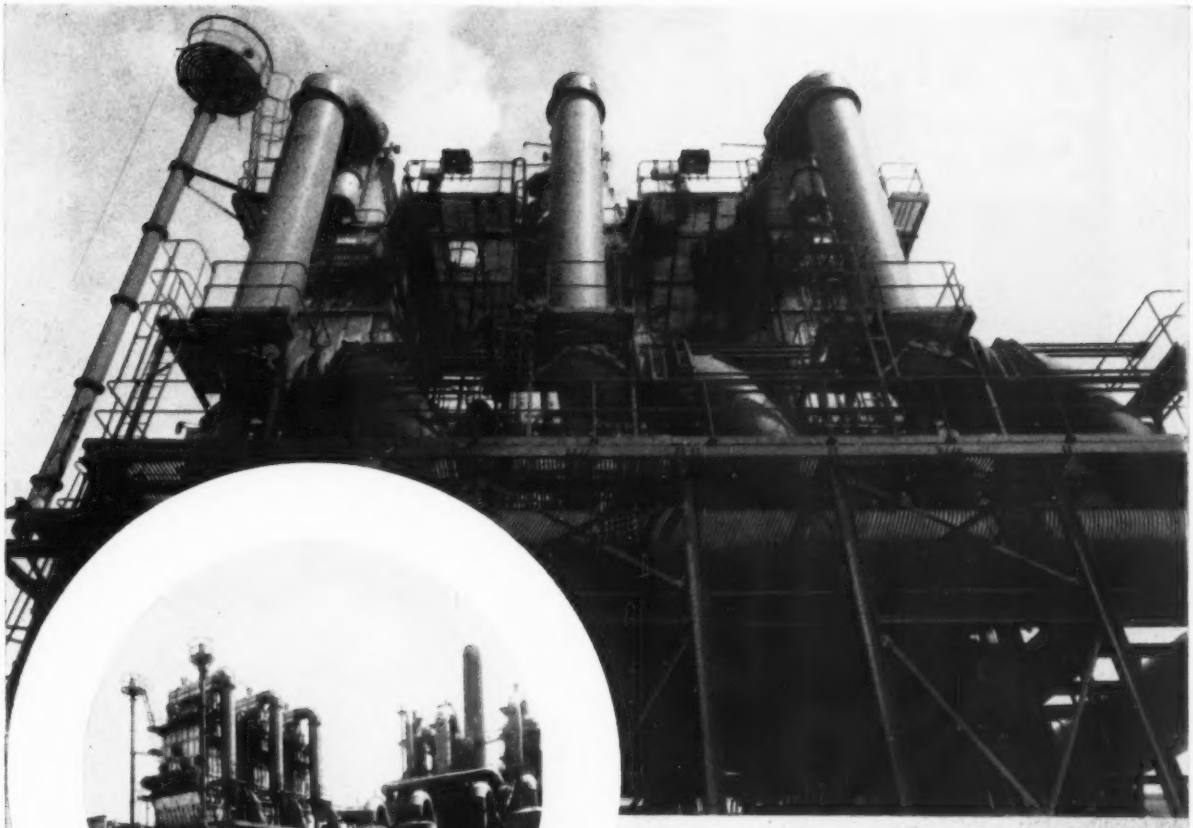

HIGH EFFICIENCY GAS CLEANING

ferro-manganese gas from blast furnaces

The photographs reproduced above show the three Holmes-Elex Electrical Precipitators cleaning ferro-manganese gas after the blast furnaces at the works of the Workington Iron & Steel Company (Branch of the United Steel Companies Limited). This installation has a total capacity of 4,200,000 c.f.h. at N.T.P.

ELECTRICAL EQUIPMENT
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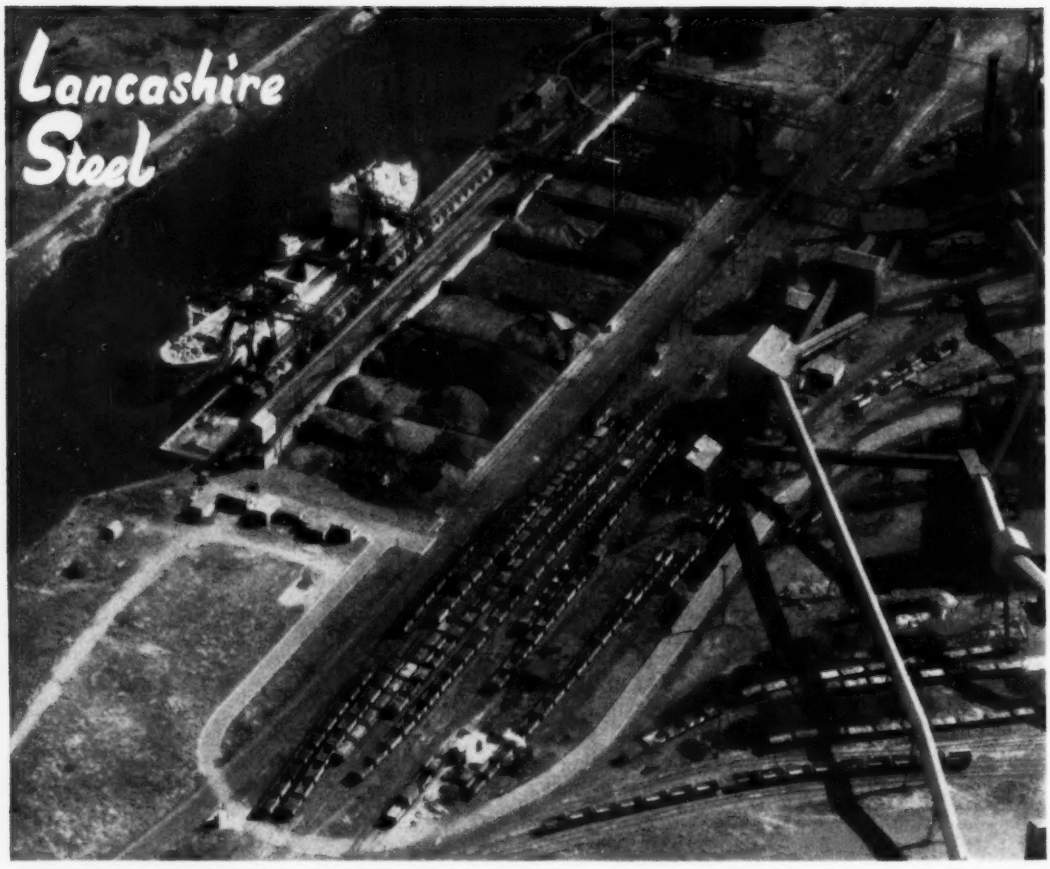

AERIAL VIEW OF THE UNLOADING WHARF—*Photograph by Aero Pictorial Ltd.*

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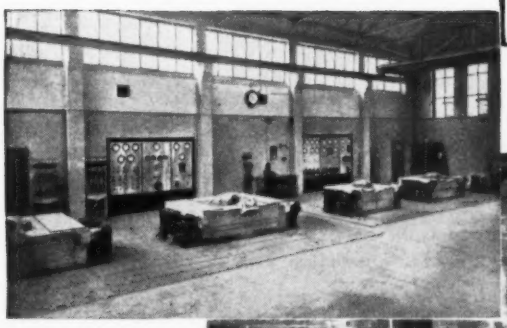
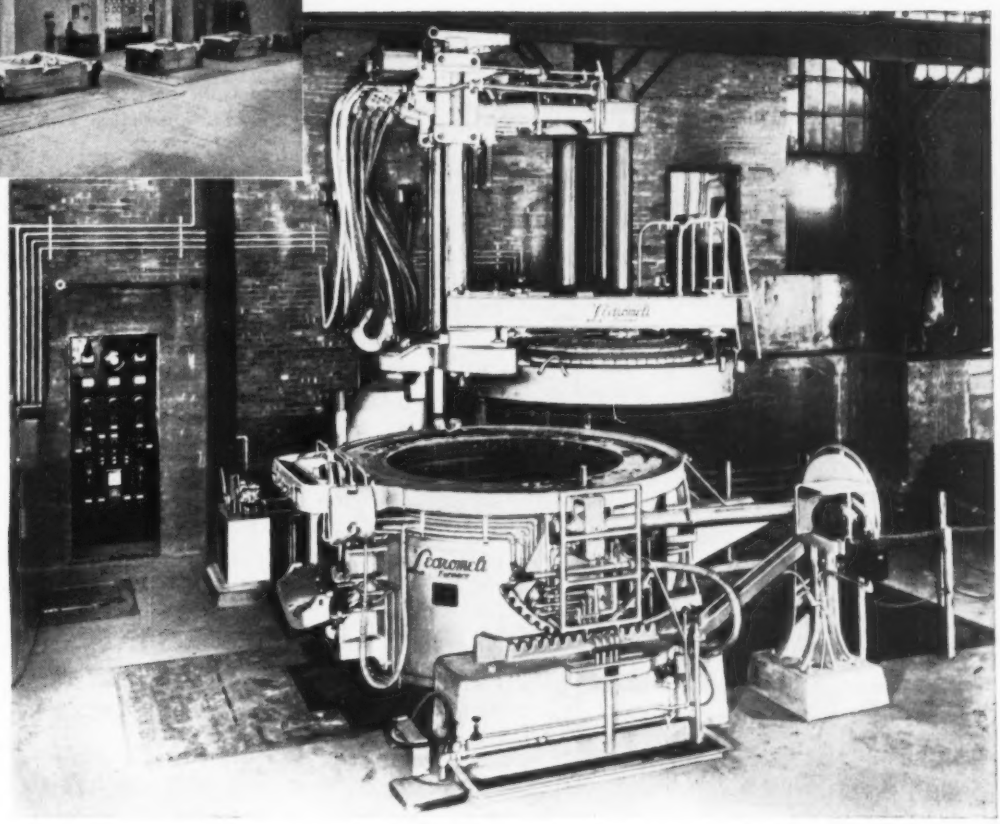
GWB

DEMAG

Lectromelt



Demag high-frequency induction furnace installation at a steel making plant, incorporating two 1 ton capacity furnaces, two 660 lb. capacity furnaces, and one 110 lb. capacity furnace.



A 3 ton capacity Lectromelt arc furnace of 9'-0" diameter shown with roof swung ready for charging.

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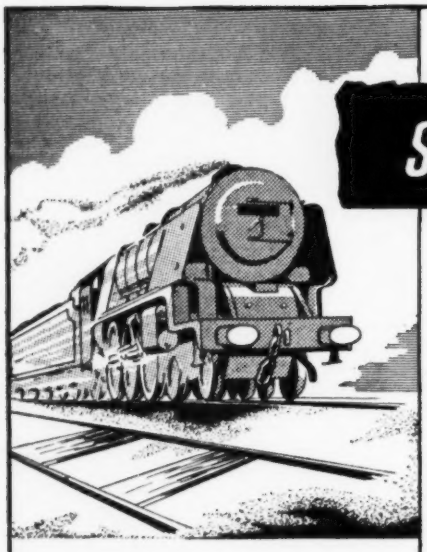

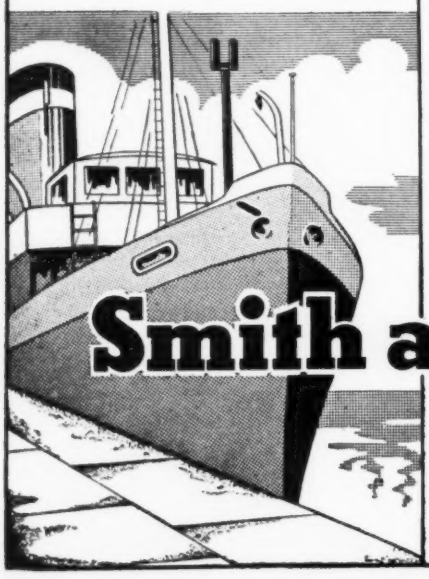

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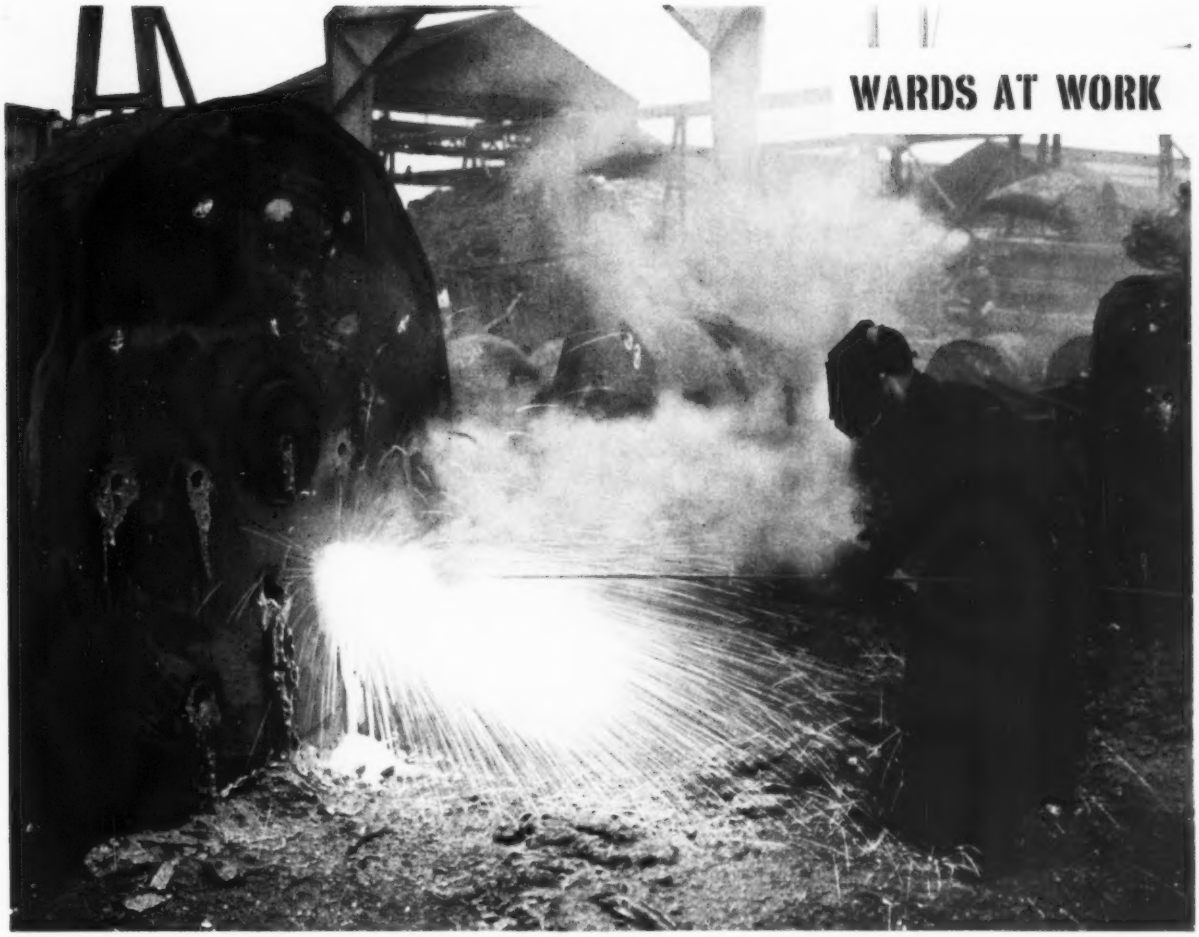

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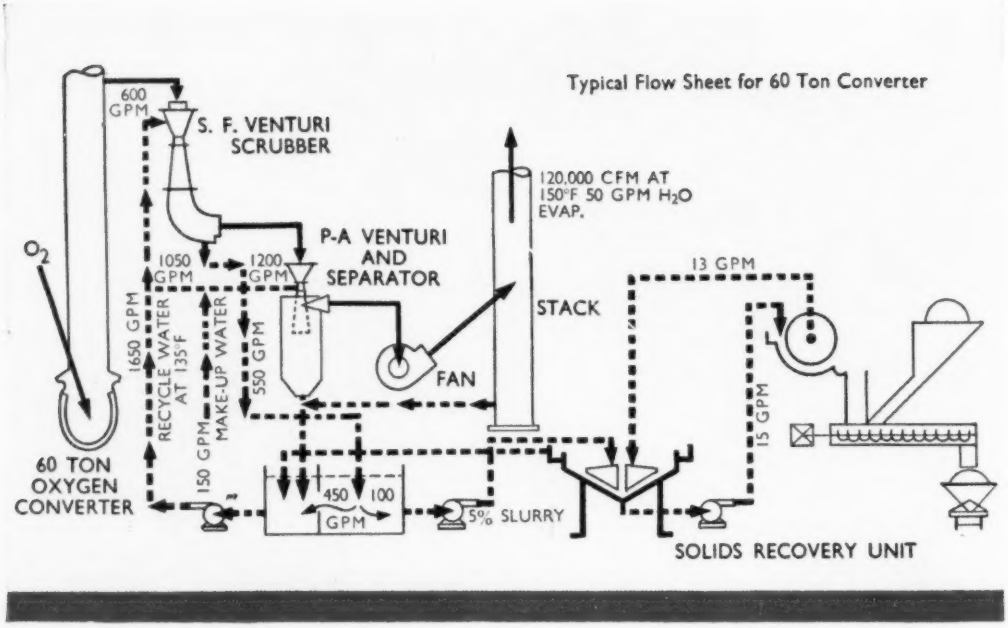

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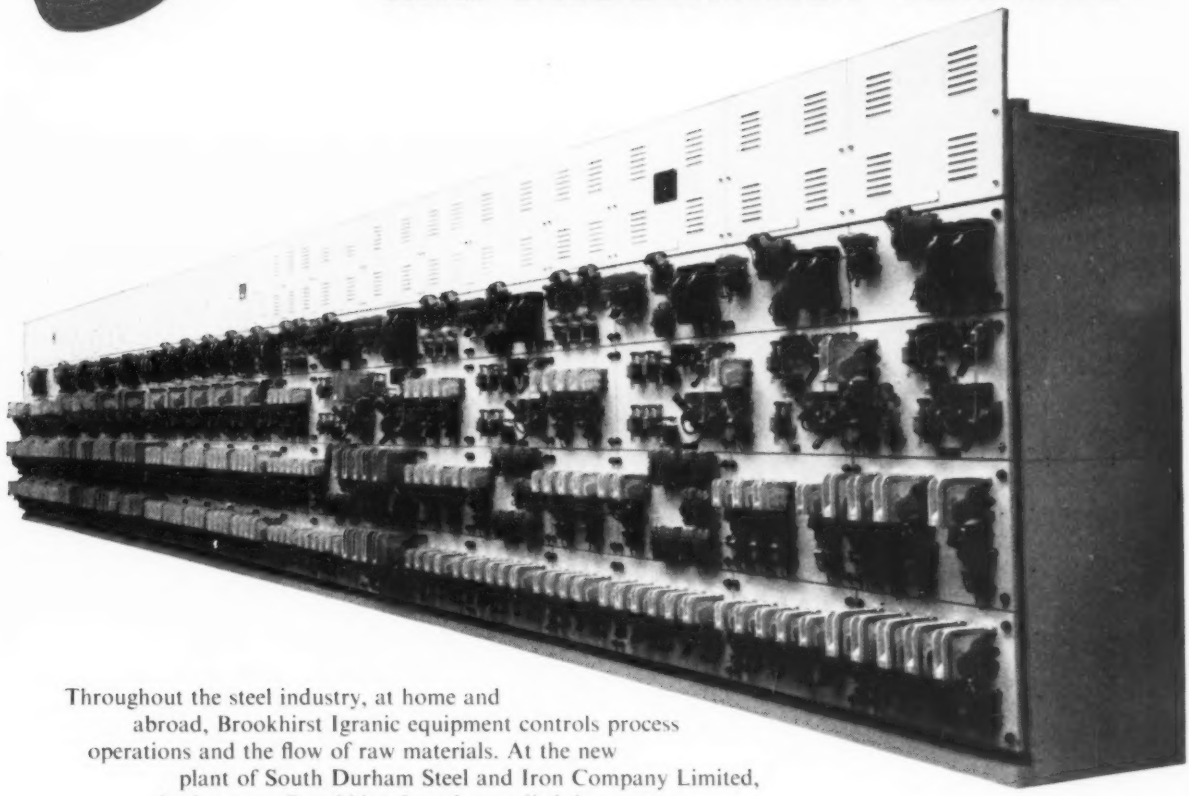

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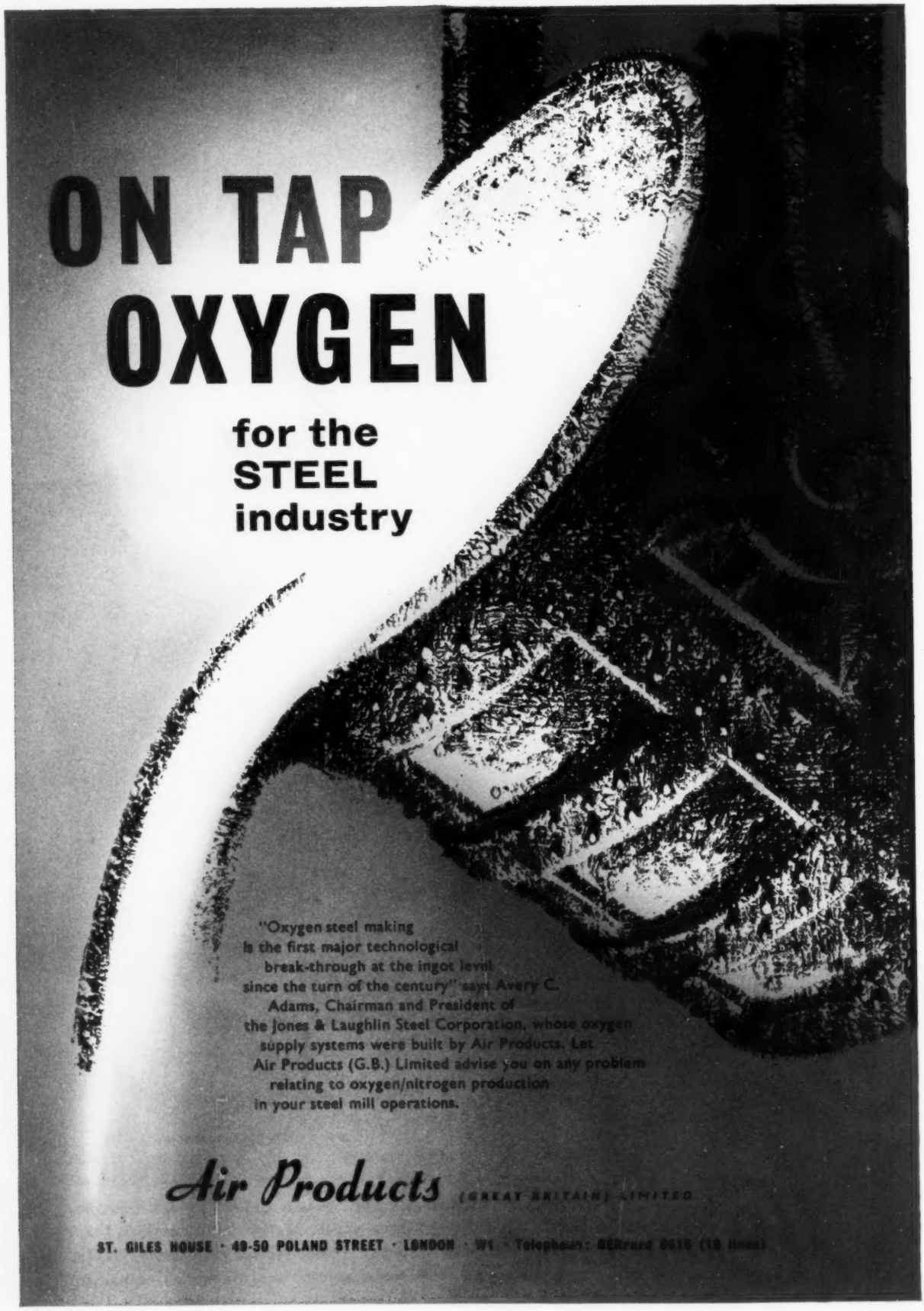
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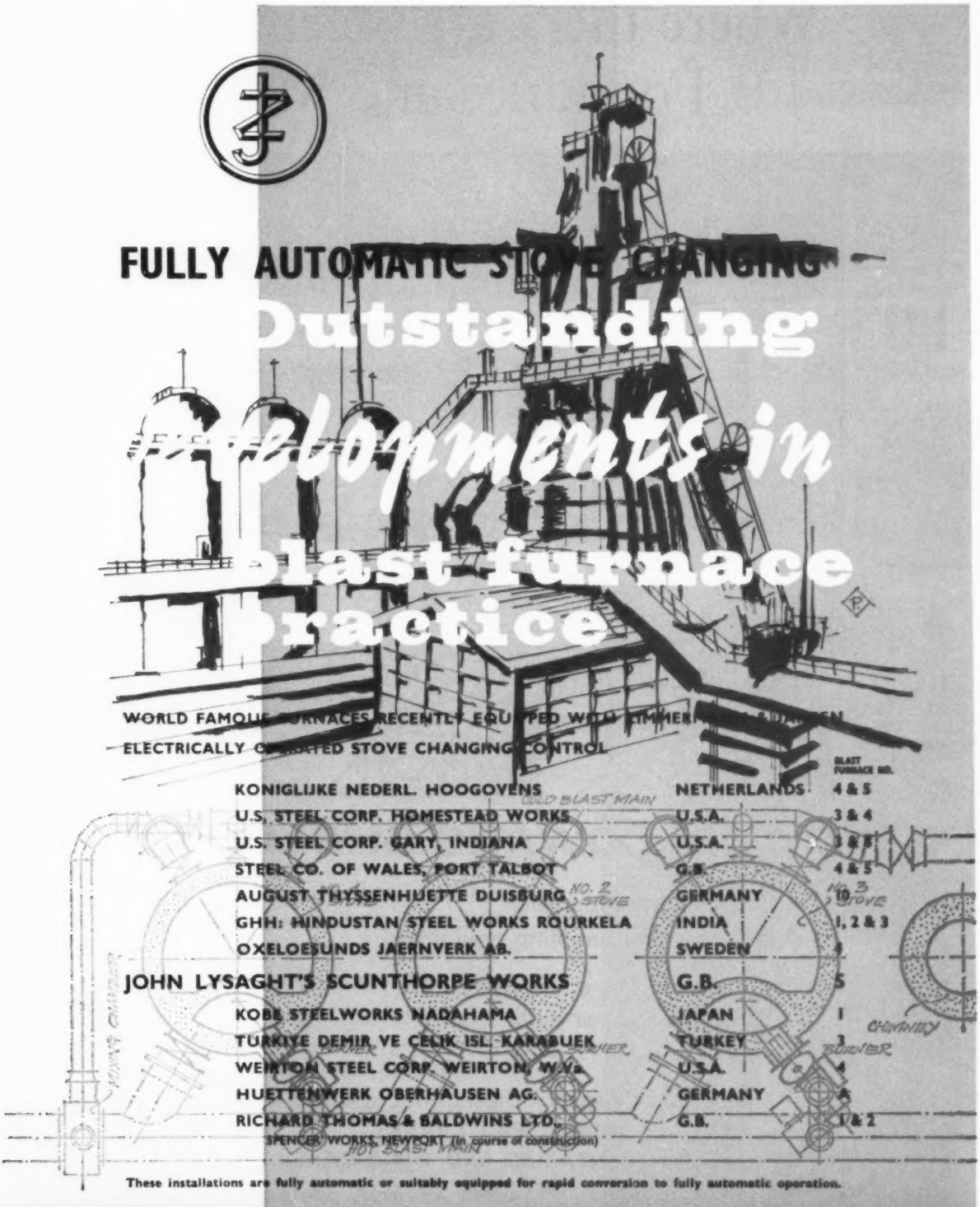

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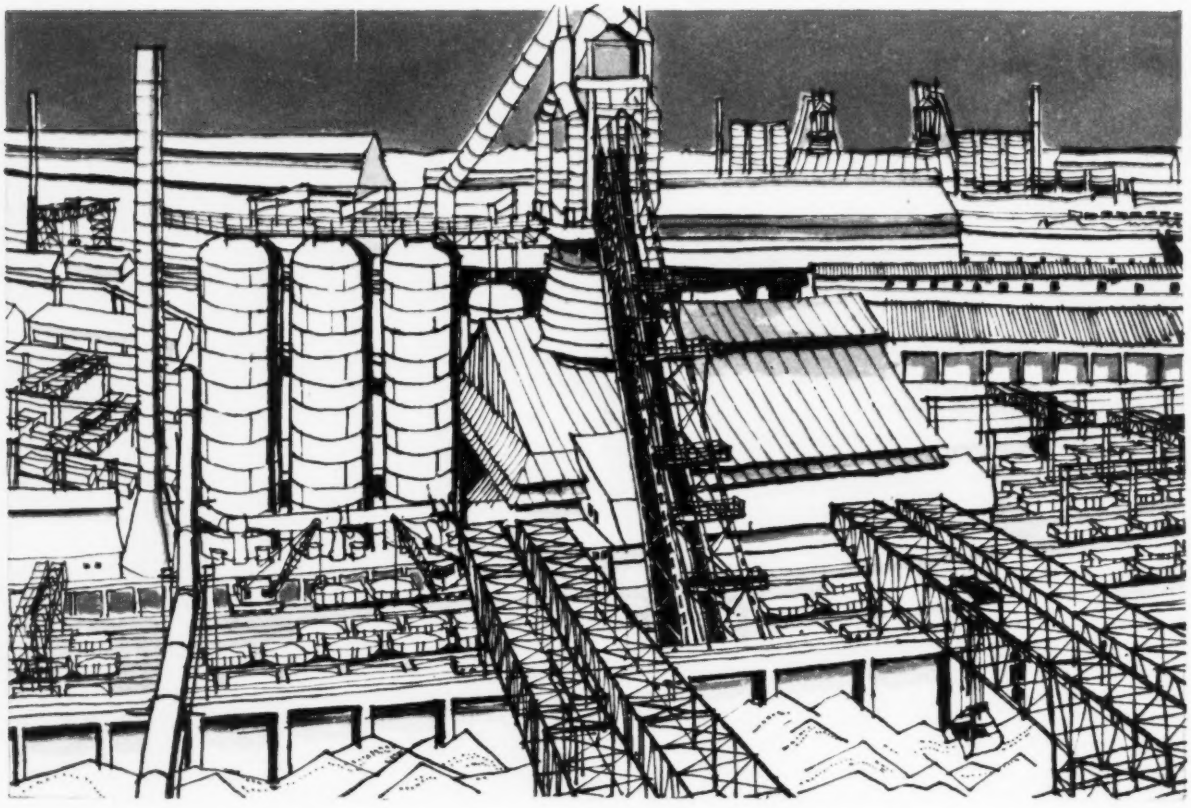

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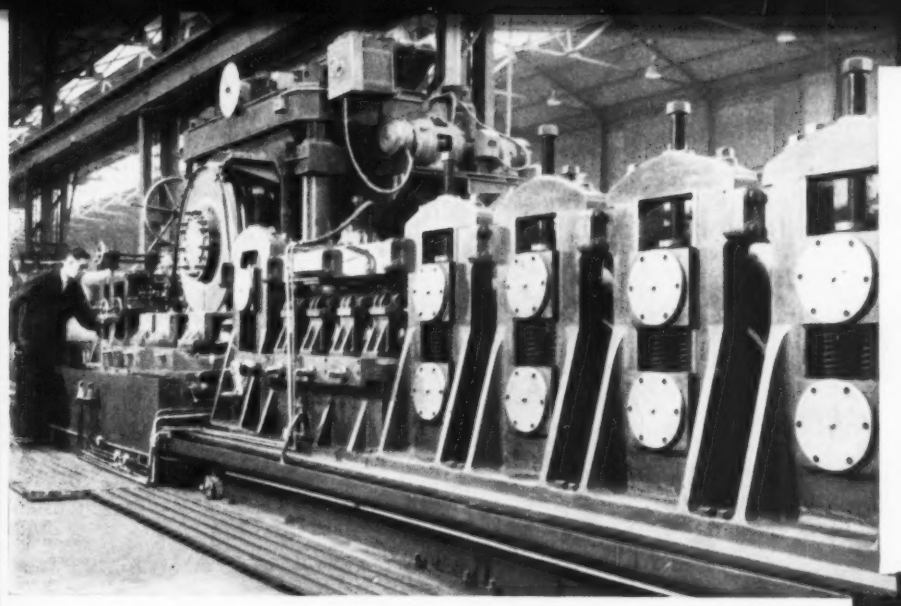

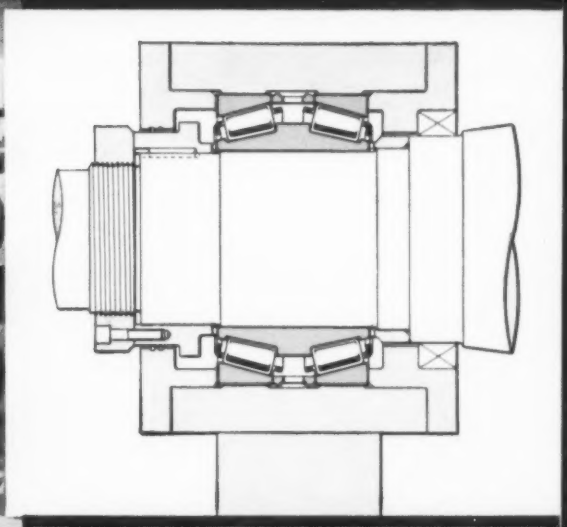

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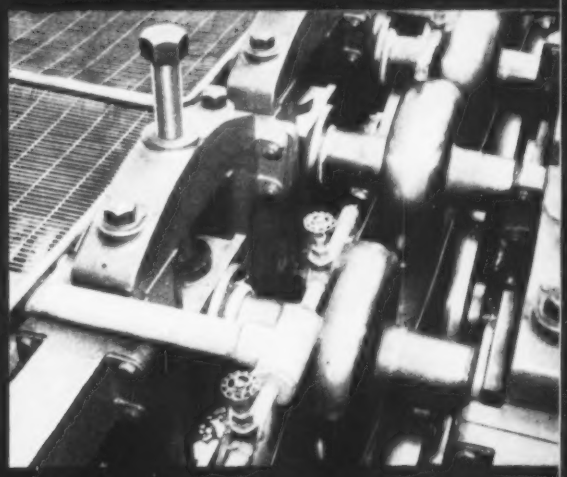
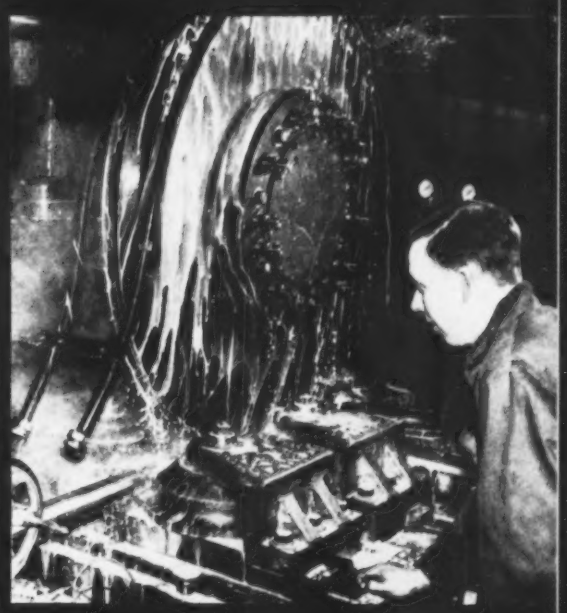
Above can be seen the forming-roll unit equipped with Timken bearings, while on the right the partially-formed strip is seen passing between the horizontal forming-rolls running in Timken bearings, mounted as shown in the drawing.

The lower view shows the welding stage of the E.R.W. process and here again the main spindle carrying the electrodes runs in Timken bearings.

British Timken, Duston, Northampton, Division of The Timken Roller Bearing Company. Timken bearings manufactured in England, Australia, Brazil, Canada, France and U.S.A.





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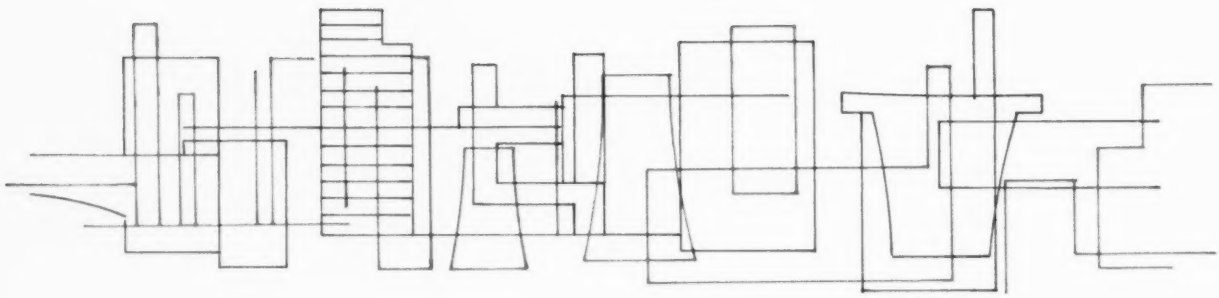

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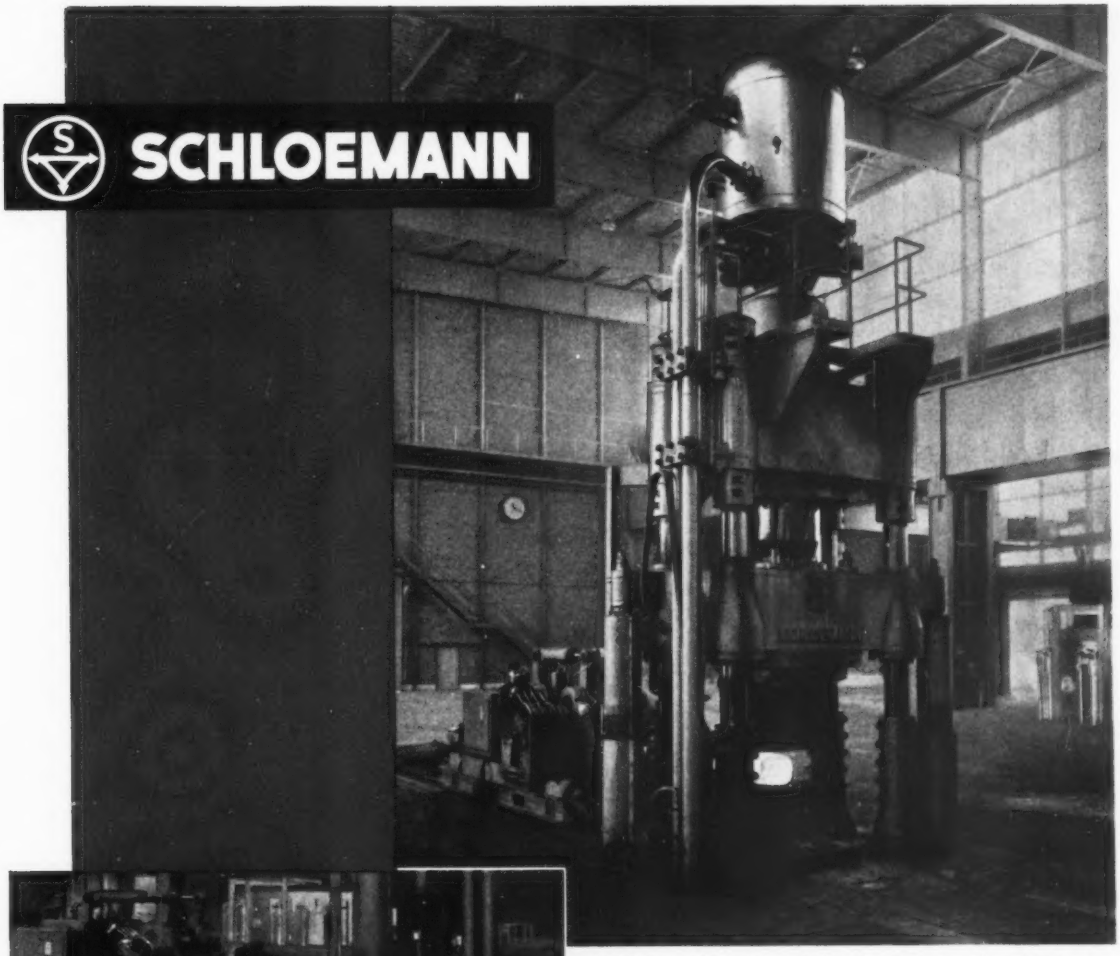

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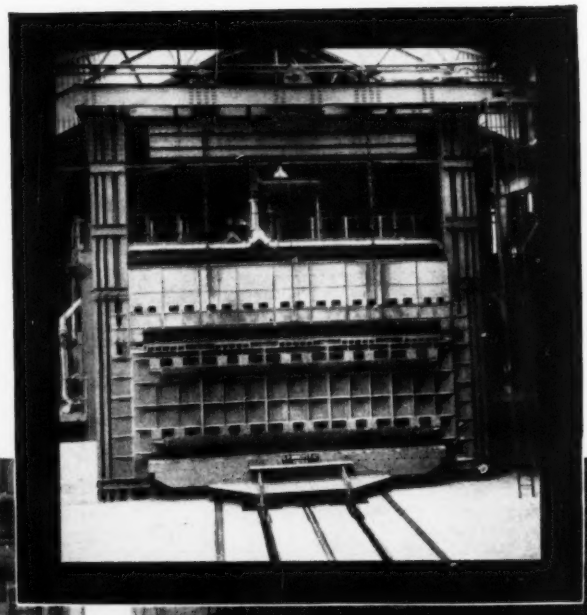

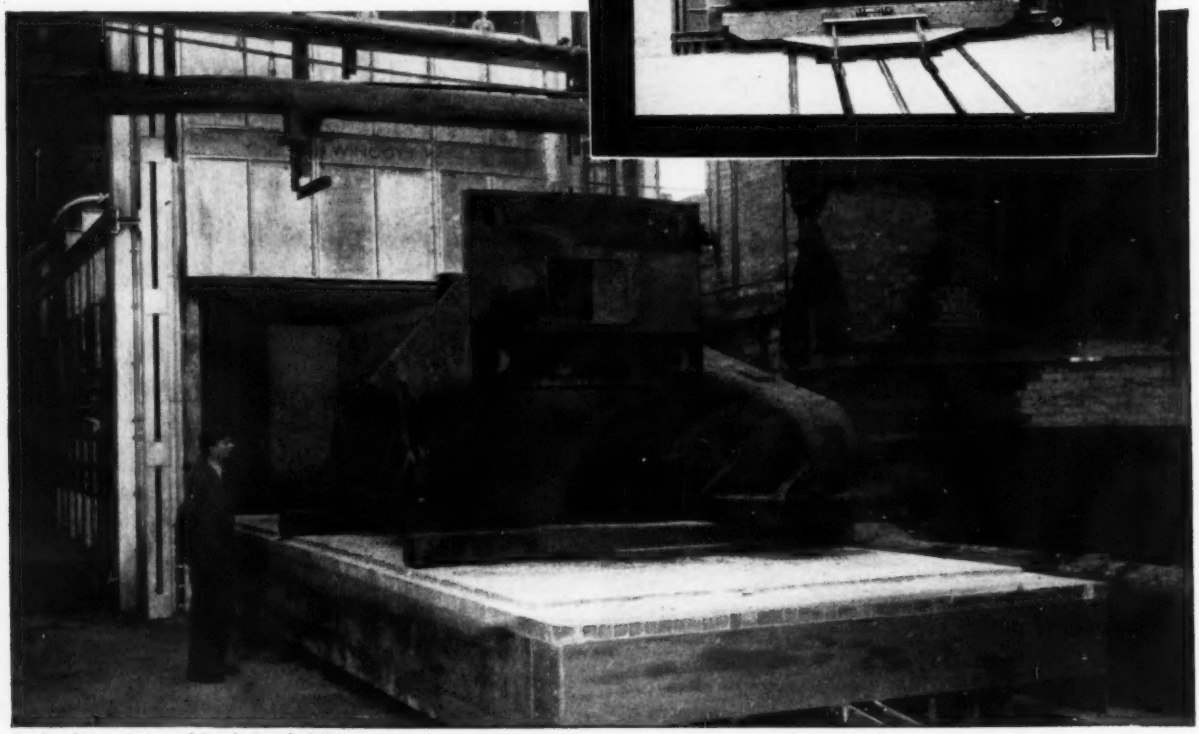

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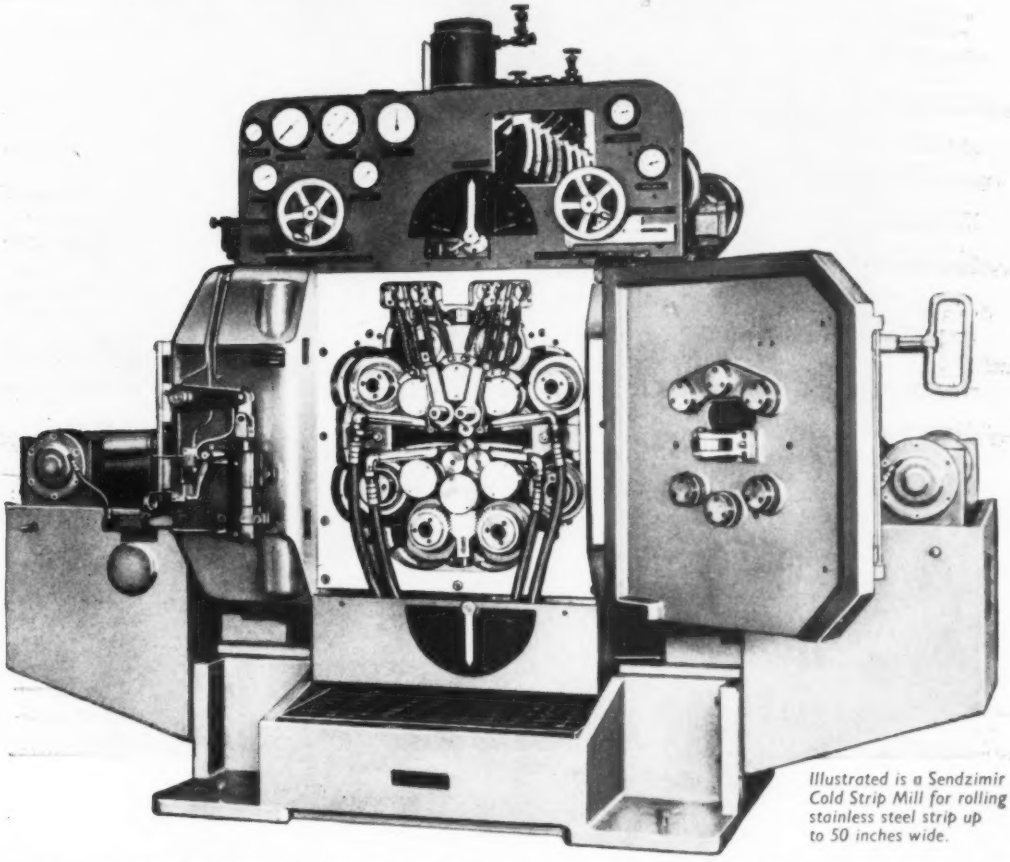

Illustrated is a Sendzimir Cold Strip Mill for rolling stainless steel strip up to 50 inches wide.

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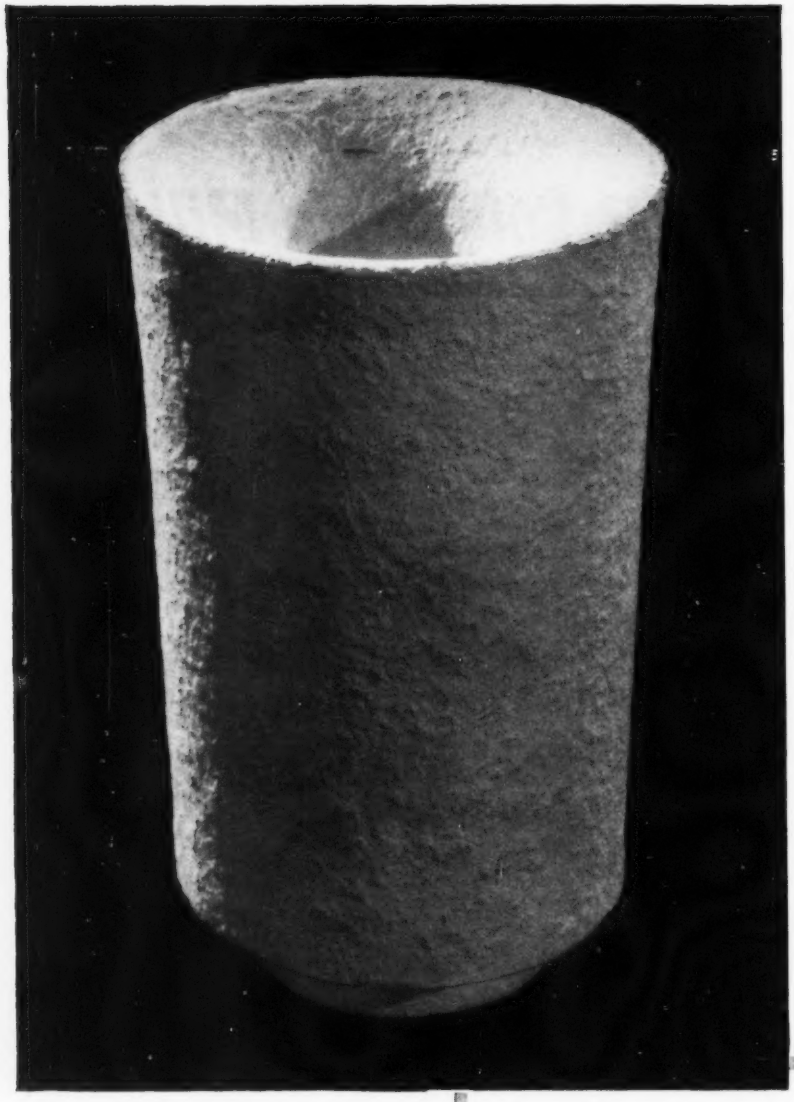

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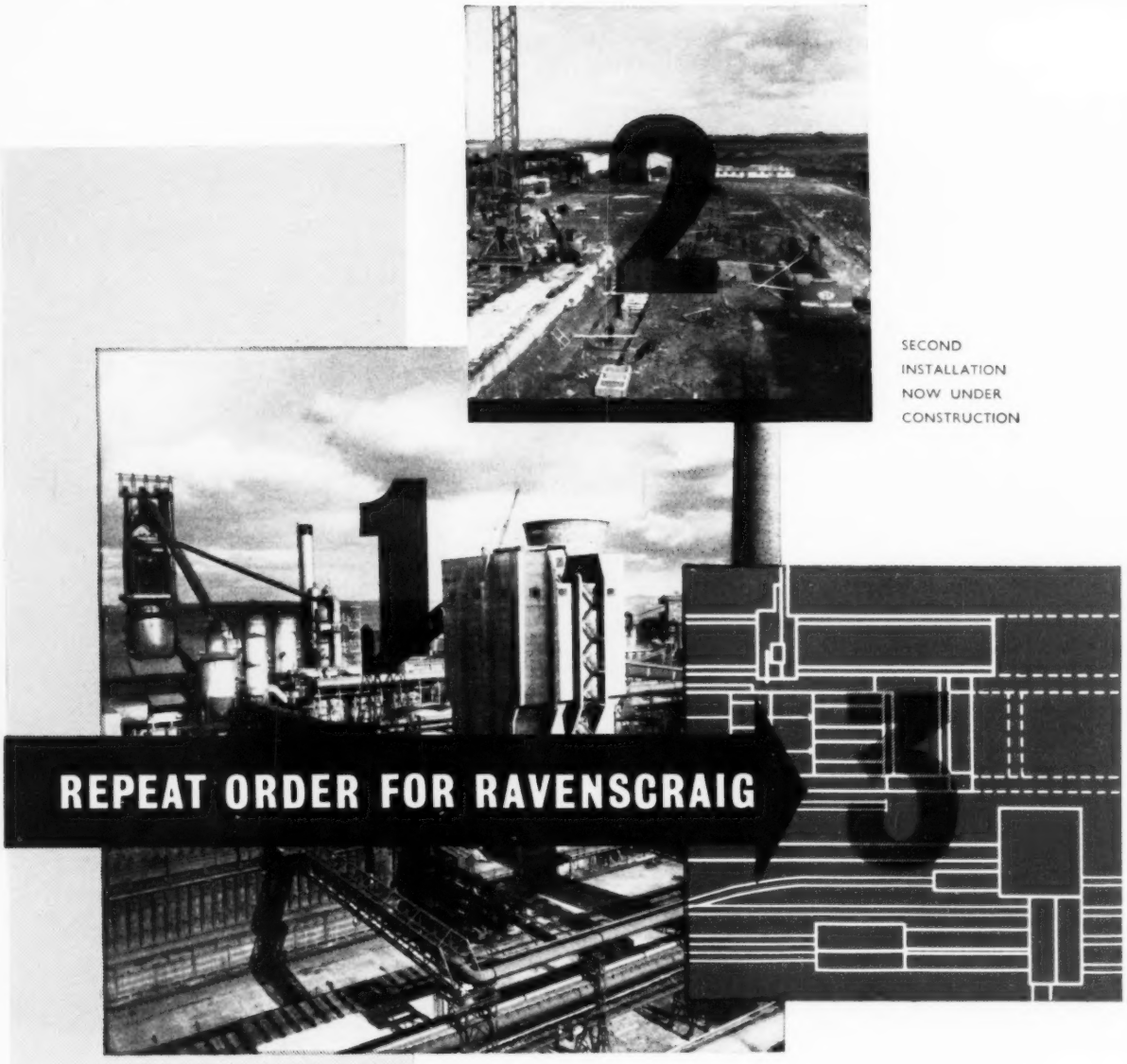

SECOND INSTALLATION NOW UNDER CONSTRUCTION

REPEAT ORDER FOR RAVENSCRAIG

Woodall-Duckham has received the order for a third installation of Coke Ovens and By-Product Plant at the Ravenscraig Works of Colvilles Limited. It will comprise two batteries of 35 ovens with complete coal and coke handling systems and an extensive indirect by-product recovery plant.

By the end of 1962 six batteries will be in operation at Ravenscraig, with a total carbonising capacity of some $1\frac{3}{4}$ million tons of coal yearly.

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Contents

THE IRON AND STEEL INSTITUTE

Report of the Annual General Meeting 1960 — 377

The adherence of oxide films on metals — 380
R. F. Tylecote

The physical metallurgy of 12% chromium steels — 386
K. J. Irvine, D. J. Crowe, and F. B. Pickering

Under-hearth cooling for blast-furnaces — 405
E. M. Summers

Effect of Al and N on the occurrence of intergranular fracture in steel castings — 409
B. C. Woodfine and A. G. Quarrell

Practical estimation of composition balance and ferrite content in stainless steels — 415
L. Pryce and K. W. Andrews

A radiochemical investigation of the combustion method for the determination of sulphur in cast iron — 417
R. C. Rooney and F. Scott

Serrations in the stress/strain curve of cold-worked 301 stainless steel at 20 K — 439
J. F. Watson and J. L. Christian

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

The notch ductility and tensile properties of some synthetic mild steels — 422
A. M. Sage and F. E. L. Copley

CORRESPONDENCE ON PAPERS

Abraham Darby and the beginnings of the coke-iron industry — 473
H. R. Schubert

DISCUSSION ON PAPERS

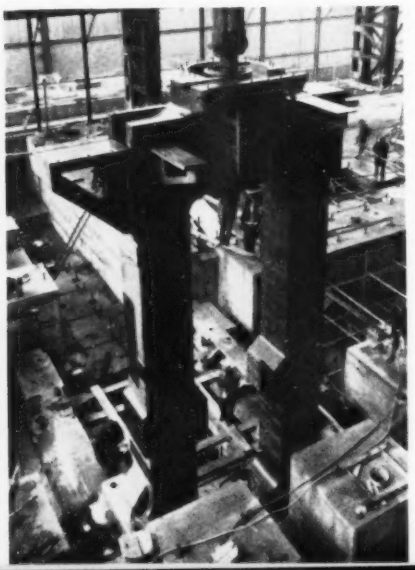
Discussion at the Autumn Meeting 1959 on:
- Forging — 440
- Carbides and nitrides — 450
- Controlled-transformation stainless steels — 457
- Fuming in steelmaking furnaces — 460
- Tempering — 466

News 475 **Translations** 478 **Abstracts** 479

THE IRON AND STEEL INSTITUTE
4 Grosvenor Gardens London SW1

STEEL PLATE FOR BRITISH INDUSTRY

Colour photograph by Studios Fifty-One Limited, by courtesy of the United Steel Companies Limited.



Of five new major plate rolling plants under the British Iron and Steel Industry's development programmes, Davy-United secured orders for three—Appleby-Frodingham Steel Company, South Durham Steel and Iron Company Limited and Consett Iron Company Limited.

These three Plants are all designed to produce high quality steel plate much in demand for Britain's shipbuilding, atomic energy and engineering industries. The first is already in production, the 39" and 63" x 114" 4-High Reversing Hot Plate Finishing Mill built for Appleby-Frodingham Steel Company. Driven by two 4,000 h.p. motors, this Mill rolls plate up to 12 ft. finished width, in thicknesses up to 4½" and maximum lengths of 65 ft.

DAVY-UNITED

Issued by the Machinery Division of
Davy and United Engineering Company Limited, Sheffield

353/DU

H. BULL, D.Met., F.I.M.

HARRY BULL was born in Sheffield in 1893 and began working in the laboratory of Wm Jessop and Sons when he was 13 years of age.
He studied at evening classes in Sheffield, and his early contacts included Arnold, Ibbotson, Brearley, and Hatfield. He was awarded an Associateship of Metallurgy at the University of Sheffield in 1918.
Dr Bull gained wide theoretical and practical experience by working with a number of Sheffield firms before joining Brown Bayley Steels Ltd in 1924 as chief chemist. In 1937 he became works manager and later works director and joint managing director. He was appointed managing director of Brown Bayley Steels in 1951, a post he resigned in April 1959 although he remains on the board.
Dr Bull's lifelong interest in education has ranged from the training of steelworks operatives to the work of university postgraduates. In recognition of this, as well as his direct contributions to the iron and steel industry, Sheffield University awarded him the honorary degree of Doctor of Metallurgy in 1959.
In 1946 Dr Bull was elected a Fellow of The Institution of Metallurgists. He has been a member of The Iron and Steel Institute since 1918, and as president of the Sheffield Society of Engineers and Metallurgists for 1959–60 he served as an Honorary member of the Council of the Institute.

H. Bull, D.Met., F.I.M.

Honorary Member of Council, 1959–60

Report of the Annual General Meeting 1960

The Ninety-first Annual General Meeting of The Iron and Steel Institute was held on Tuesday, Wednesday, and Thursday, 3, 4, and 5 May 1960, at The Royal Commonwealth Society, Northumberland Avenue, London WC2; Caxton Hall, Caxton Street, London SW1, and Church House, Great Smith Street, London SW1. Mr W. Barr, O.B.E., *the Retiring President, was in the Chair at the beginning of the Meeting, his place being taken later by Mr W. F. Cartwright,* D.L., J.P., *the new President.*

Minutes of previous Meeting

The minutes of the previous Meeting held on 2 December 1959, were taken as read, and were confirmed and signed.

Welcome to members and visitors

The **President,** on behalf of the Council, extended a very hearty welcome to Members and visitors. There were many distinguished members and visitors from abroad. The Institute was delighted that they had honoured it by coming to the Meeting and hoped that they would enjoy their visit.

Obituary

The **President** announced the names of Members who had died. He referred in particular to the passing of Dr C. A. Edwards, F.R.S., who had died on 29 March 1960. Dr Edwards had first become a Member of the Institute in 1907, and had served on the Council from 1937 to 1946. He had been Professor of Metallurgy at University College, Swansea, from 1920 until he retired in 1947, and Principal of the College during the last of those 20 years. In 1930 he had been elected a Fellow of the Royal Society.

The Meeting, at the invitation of the President, stood in silence for a few moments as a token of respect.

Presentation of Report of Council and Accounts

The **President** then presented the Report of Council for 1959 and the Statement of Accounts (*JISI*, 1960, **195,** May, 1–12) and moved their adoption.

Membership had continued to increase. With a total of 5204 it was over 100 more than the previous year.

Successful Annual and Autumn Meetings had been held in London. The Special Meeting on the north-east coast in July had been memorable for excellent organization and sumptuous hospitality.

The 12th Hatfield Memorial Lecture had been given in Sheffield by Academician G. V. Kurdjumov (Central Research Institute for Ferrous Metallurgy, Moscow). The Engineers' Group had held a series of interesting meetings, including visits to Holland in April and to Glasgow in October. The Powder Metallurgy Joint Group had held two excellent meetings in London; they were grateful to the Secretary of The Institute of Metals for organizing those on behalf of the Joint Committee. An interesting meeting had been held in September to commemorate the 250th anniversary of the successful use of coke in ironmaking by Abraham Darby at Coalbrookdale. The Affiliated Local Societies had been very active and a number of good joint meetings had been arranged.

The *Journal* had been published as usual. The printing dispute in the summer had caused some delay which had not yet been entirely overcome. The *Journal* would be larger than usual during 1960, because papers and abstracts were to be published in the shortest practicable time. This was estimated to involve extra expenditure of nearly £4000.

In 1959, by arrangement with the Department of Scientific and Industrial Research, the monthly publication of an English translation of the Russian Journal *Stal* had been undertaken. This had proved an immediate and justifiable success.

The British Iron and Steel Industry Translation Service had continued to be very active, and again nearly 400 translations had been made available at very reasonable prices.

This year a new Abstract and Book Title Index Card Service had been started, which would enable abstracts, each on a card, to be made available to subscribers much more quickly than they could be printed in the *Journal*. The number of subscriptions already received indicated that this service was likely to be of considerable value.

The Joint Library and Information Department had been operating under considerable difficulties, owing to the disturbances caused by building operations. The extensions at 4 Grosvenor Gardens were now finished and badly needed additional accommodation had been made available for staff and for storing books and periodicals. The Reading Room had been restored for the use of Members. Increased use had been made of the Library and Information Department. The President expressed the Council's appreciation of the work of Mr M. L. Pearl, Librarian, Mr J. P. Saville, Information Officer, Miss C. L. Fallowfield-Cooper, Senior Library Assistant, and all the other members of the staff during the disturbance.

As in other years, most friendly relations had been maintained with many other societies, both in the UK and overseas. Close association with the British Iron and Steel Research Association, the British Iron and Steel Federation, the Affiliated Local Societies, The Institute of Metals, and The Institution of Metallurgists had been continued. Some of the Council thought that even closer co-operation among societies which catered for the needs of metallurgists would be profitable.

During the year the Newport and District Metallurgical Society, which had attained its 25th birthday, decided to take

steps to establish a permanent memorial to Sidney Gilchrist Thomas. It was curious and deplorable that there was no sufficient memorial to Thomas and to his cousin, Percy Carlisle Gilchrist, in the UK. The steel industry of the world owed these great men an overwhelming debt. By their invention of the basic lining they had liberated for profitable use the major phosphoric ore deposits of the world. The Council had therefore been pleased to support the initiative of the Newport Society. A granite obelisk was being erected in South Wales near the site where the first works experiments with a converter had taken place, and it was proposed to set up a scheme of travelling scholarships. Some companies had contributed generous sums of £500 and £1000 each. He hoped that others would follow suit, and that this appeal would have the support of the steel industry in Europe, in the USA, and elsewhere. The German steel industry had made a very generous contribution of 5000 marks (over £4000) and he took this opportunity of thanking them publicly.

The **Honorary Treasurer** (Sir Julian Pode) presented the Accounts and seconded the President's motion. At the Meeting last year the President had indicated that in 1960 there might be a deficiency of £4000. In the event, it was considerably less. From 1 January 1960, increased subscriptions had become payable, and as a result it was estimated that there would be a small surplus at the end of the financial year 1960, amounting to £2000, plus £1000 contingencies.

As was shown in the Accounts, from 1 January 1960, a more modern superannuation scheme had been arranged for the staff, and this would cause additional expense which had been taken into account in the estimate which he had quoted for the future year.

Sundry creditors and debtors showed a substantial increase, the result of delays caused by the dispute in the printing industry during 1959.

The motion for the adoption of the Report of Council and the Statement of Accounts was carried unanimously.

Presentations

Bessemer Gold Medal

The **President** announced that the Bessemer Gold Medal for 1960 had been awarded to Professor Dr Hermann Schenck of Germany, in recognition of his distinguished contributions to metallurgical research, and with special reference to his pioneer investigations into the physical chemistry of steelmaking.

Professor Schenck, after a distinguished early career, was now Professor of Metallurgy at the Technische Hochschule, Aachen, and since 1950 he had been President of the Verein Deutscher Eisenhüttenleute. Among his many published scientific and technical works, first place should be given to his book 'The introduction to the physical chemistry of steelmaking', written in 1932–4. This was the first book ever to deal with the problems of liquid iron and steel and their interactions with slag, and thus Professor Schenck had been one of the first to try to synthesize the whole field of steelmaking reactions from the thermal–dynamic viewpoint. The book had been translated into English and Italian. The English edition, translated by Dr Goldschmidt, was published in 1945 by BISRA.

Professor Schenck's contributions and services to the iron and steel industry in many parts of the world had been widely recognized. He was an Honorary Member of the American Society for Metals, the American Institute of Mining and Metallurgical Engineers, the Iron and Steel Institute of Japan, and the Société Française de Métallurgie. He had been awarded an Honorary Doctorate in 1958 by the Berlin Technical University, and in 1959 the Société Française de Métallurgie had presented to him its Grande Médaille. This year it was the Institute's turn to honour Professor Schenck, to whom the President presented the Bessemer Medal.

Professor Dr **Hermann Schenck** thanked the Council for the honour conferred upon him and conveyed to them the greetings and good wishes of their German colleagues in the Verein Deutscher Eisenhüttenleute. For many years the two societies had been working together to attain the same objective, the exchange of experience and knowledge which results from dealing with iron and steel. Unless geographical and political frontiers were transcended by scientific discussion of problems, great progress would not be possible. Collaboration between scientists was often more unselfish than that between politicians, and contributed to the well-being and security of the world. Common interests had led to friendship between The Iron and Steel Institute and his own Society.

He deeply appreciated the very high honour and offered his warmest thanks to the council of the Institute.

The Sir Robert Hadfield Medal

The **President** announced that the Sir Robert Hadfield Medal for 1960 had been awarded to Dr J. C. Hudson, of the British Iron and Steel Research Association, in recognition of his long and distinguished work in the field of corrosion of iron and steel.

Dr Hudson was well known as head of the BISRA Corrosion Section, and this was the second major award bestowed upon him within the past six months. In November 1959, the American National Society of Corrosion Engineers had given him an award in recognition of public contributions to the field of corrosion engineering.

Dr Hudson had had a distinguished career. After obtaining a degree and Associateship at the Royal College of Science, he had undertaken research in physical chemistry. This had led to his appointment as investigator to the Atmospheric Sub-Committee of the British Non-Ferrous Metals Research Association. The results of that work had been published by the Faraday Society and formed the basis of a thesis for which he had been awarded the D.Sc. degree of London University. Then he became investigator to the newly formed Corrosion Committee of The Iron and Steel Institute, and had continued in that post when the Committee was taken over by BISRA 12 years ago. From 1953 to 1955 Dr Hudson had been Chairman of the Corrosion Group of the Society of Chemical Industry.

Dr **J. C. Hudson,** expressing his thanks to the President and the Council of the Institute, said that he greatly appreciated the honour that had been paid him. It was perhaps not inappropriate, in view of Sir Robert Hadfield's interest in the corrosion field, that the medal should have been awarded to a corrosion worker. He regarded this as a tribute not only to himself, but also to the colleagues who had worked with him, the hundreds of people, including the President, who had served on the Corrosion Committee and who had given him the utmost possible assistance.

A Williams Prize for 1959

The **President** presented a Williams Prize for 1959 of £100 to Mr I. M. D. Halliday (United Steel Companies Ltd) for his paper on 'Continuous casting at Barrow' (*JISI*, 1959, **191,** 121–163).

Mr **I. M. D. Halliday** expressed his thanks for the great honour, which he regarded as a tribute to his co-workers, the staff at Barrow, and the men who actually worked on that plant.

Ablett Prizes for 1959

The **President** presented Ablett Prizes for 1959 of £100 each to Mr V. Harrison and Mr V. Stephenson (Dorman Long (Steel) Ltd) for their paper on 'The new medium-section mill at the Cleveland works of Dorman Long (Steel) Ltd' (*JISI*, 1959, **192,** 132–142).

Announcements

Andrew Carnegie Silver Medal

The **President** announced the award of an Andrew Carnegie Silver Medal for 1959 to Dr P. R. V. Evans (formerly University College, Swansea, and now at Armour Research Foundation, USA) for a report on 'The effect of rolling unstable austenitic 0·76% carbon steel at 220–300°C' (*JISI*,

1959, **191,** 34–44). The co-author, Professor H. O'Neill, was not eligible. Dr Evans was unable to be present at the Meeting.

Retiring Members of Council

In accordance with bye-law 10, the names of the following Vice-Presidents and Members of Council had been announced at the Autumn Meeting 1959, as being due to retire at the present Annual Meeting:

Vice-Presidents: Mr D. A. Oliver, C.B.E., Mr N. H. Rollason, and Sir Charles Goodeve, O.B.E., F.R.S.

Members of Council: Dr J. H. Chesters, Mr F. B. George, Mr W. F. Gilbertson, Dr L. B. Pfeil, O.B.E., F.R.S., and Mr R. F. Summers.

No other members having been nominated up to one month previous to the present Meeting, the retiring Vice-Presidents and Members of Council were presented for re-election.

Changes on the Council

Since the last General Meeting on 2 December 1959:

Honorary Vice-President: Mr G. H. Johnson

Member of Council: Mr P. E. Holloway

Honorary Members of Council:

Sir Ronald Prain, O.B.E., President of The Institute of Metals, in place of Mr G. L. Bailey, C.B.E.

Mr W. L. Bolton, President of the Leeds Metallurgical Society, to succeed Mr K. Hill.

Mr R. Cobill, President of the Liverpool Metallurgical Society, to succeed Mr B. Ryan.

Mr W. C. Heselwood, President of the Sheffield Metallurgical Association, to succeed Professor A. G. Quarrell (Member of Council).

Professor R. J. Sarjant, President of the Sheffield Society of Engineers and Metallurgists, to succeed Dr H. Bull.

Additional Honorary Members of Council

Dr J. Nutting.

Dr W. S. Walker.

Election of Members

The scrutineers, Professor **F. D. Richardson,** and Mr **A. N. Whiting,** reported that all the candidates for membership had been duly elected. They numbered 132 Members and 46 Associate Members. The total membership was now 5204.

The President declared the applicants duly elected.

Meeting in Italy

The **President** announced the following elections:

To be an Honorary Member of the Institute: Prof. Dott. Felice De Carli, President of the Italian Metallurgical Association.

To be Honorary Vice-Presidents of the Institute: Dott. Ing. Giovanni Falck, Chairman and Managing Director of the Falck Company; Prof. Ernesto Manuelli, Chairman and Managing Director of the Finsider Company.

At the Joint Meeting in Milan on 30 May his successor would present the diplomas to them.

The **President** then announced that the Italian Metallurgical Association had awarded the *Federico Giolitti Steel Medal* to Mr W. F. Cartwright. This was indeed a great honour. The Medal, which was recently founded, had only been awarded once before, in 1958, and that had been posthumously in memory of Giorgio Enrico Falck and of Oscar Sinigaglia. The presentation would also be made in Milan.

Induction of new President

The retiring **President** (Mr W. Barr, O.B.E.), before inducting the President-Elect into the Chair, spoke of his distinguished career and varied achievements. He was a member of many technical and scientific organizations and devoted much of his time to public service, especially in the educational field. His outstanding contributions to the iron and steel industry in the UK had been recognized by the award of the Bessemer Medal in 1958.

Vote of thanks to the retiring President

The new **President** (Mr W. F. Cartwright, D.L., J.P.), proposing a vote of thanks to Mr Barr for his services during his year of office, said that Mr Barr had served the Institute previously for six years as the Honorary Treasurer. From the announcements it was clear that his term of office had been a year of achievement.

The vote of thanks was seconded by the Hon. **R. G. Lyttelton** (Past-President) and carried with acclamation.

Presidential address

The President then delivered his address 'Production control in an integrated iron- and steelworks' (*JISI*, 1960, **195,** June, 137–144).

A vote of thanks to the President for his address, proposed by Mr **E. T. Judge** (Vice-President) was carried with acclamation.

Technical Sessions

A number of papers were presented and discussed at the technical sessions on all three days of the Meeting. Details were given in the News section of the February issue of the *Journal* (p.249). Reports of the discussions have been published in the *Journal* for July and August of this year.

Dinner for Members

The Annual Dinner for Members was held at Grosvenor House, Park Lane, London W1, on the evening of Wednesday 4 May, and was attended by about 1500 Members and guests. The toast of *The Iron and Steel Institute and Industry* was proposed by the Rt Hon. Richard Wood, P.C., M.P., Minister of Power; the President replied. The health of the guests was proposed by Mr I. F. L. Elliot (Hon. Vice-President of The Iron and Steel Institute) and was replied to by Sir Cyril Musgrave, K.C.B. (Chairman of The Iron and Steel Board).

The adherence of oxide films on metals

A REVIEW OF INFORMATION

R. F. Tylecote

SYNOPSIS

The subject of oxide film adherence has been reviewed because of its importance in the field of high temperature materials. The possible effect of volume ratio, surface configuration, and film plasticity have been considered. Two other factors, the diffusion mechanism and the relative expansion coefficients of metal and oxide, have been considered important. In the absence of plasticity it is felt that adherence at high temperatures is primarily due to a balance between inward and outward diffusion of metal and oxygen. In non-plastic films differential contraction on cooling can result in the spalling of otherwise adherent films. 1811

INTRODUCTION

THE ADHERENCE OF OXIDE FILMS on metals is a subject seldom discussed in the now voluminous literature on oxidation. Yet it has great technical importance, since oxide films which are found to be protective at high temperatures over long periods during oxidation research, are not likely to be of much value industrially if this aspect is neglected.

This paper discusses the relationships thought likely to have a bearing on adherence, and directs attention to those apsects that seem important and worthy of further investigation. It has been restricted to pure metals and the simpler alloys and atmospheres because it was felt that our knowledge of the adherence of scales on complex alloys in industrial atmospheres had not reached a stage where any useful assessment of the various factors could be made.

The volume ratio

The volume ratio, sometimes known as the Pilling–Bedworth ratio, is the ratio of the volume of oxide to the volume of metal from which it forms. Originally Pilling and Bedworth[1] suggested that in cases where this ratio exceeded unity the films formed under compression and were protective, whereas if the ratio was less than unity the film cracked due to tensile stresses, and non-protective or linear oxidation occurred. More recent work[2, 3] has shown that this is clearly an oversimplification since oxide films on magnesium (volume ratio < 1) can be protective under certain conditions, and films formed on some metals with ratios exceeding unity break down giving rise to various types of irregular growth.[4, 5]

Furthermore, it is not altogether certain that oxides with a volume ratio greater than unity form internally stressed films in all cases. It would be possible for an oxide that grew by outward migration of metal ions to form stress-free layers on the outer surface in the later stages of growth. However, it is difficult to see how this can be done in the early stages of film formation when the oxide film is formed by a growth of oxide nuclei within the metal.

In the case of oxides in which the volume ratio exceeds unity it would be expected that any effect of the volume ratio on adherence would be more apparent in oxides such as those on V, Sb, Ta, Bi, Mo, and W, where the volume ratios are between 2 and 4, as shown in Table I. In such cases very high compressive stresses might exist.

Taking the example of tungsten, we find that a well-adhering oxide film grows steadily up to 700°C.[6, 7] Oxidation is parabolic in dry air,[8] probably forming WO_2 in contact with the metal surface with a superficial WO_3 layer above. While WO_2 probably grows by inward oxygen diffusion, WO_3 almost certainly does so, and becomes powdery above 700°C. Here then, we have two factors both militating against adherence, i.e. very large volume ratio and inward oxygen diffusion, and yet adherence appears to be satisfactory up to 700°C.

Vanadium and molybdenum both form volatile oxides at high temperatures and therefore cannot form very thick films. MoO_2 and MoO_3 form on molybdenum and adhere well up to 600°C. MoO_3 melts at 795°C and V_2O_5 at 675°C.

Very little seems to be known about the oxidation behaviour of antimony, and the melting point of bismuth is too low for useful data. Tantalum, however, forms very adherent films at low temperatures,[9] but no information is given by Kubaschewski and Schneider[10] on the adherence of tantalum after oxidation at 1250°C. Niobium at this temperature forms very adherent films which can be spalled by impact

Manuscript received 6 January 1960.
The author is at the Department of Metallurgy, University of Durham, King's College, Newcastle upon Tyne.

TABLE I Volume ratios

Metal	Oxide	Volume ratio
W	WO_2	2·08
W	WO_3	3·40
V	V_2O_3	3·18
Ta	Ta_2O_5	2·33
Nb	Nb_2O_5	2·84
Bi	Bi_2O_3	2·27
Mo	MoO_3	3·4
Sb	Sb_2O_5	2·35

after cooling to room temperature.[10] However, at lower temperatures there is a transformation from the metastable α form to the stable β form, which can result in a non-coherent film.[11]

Thus there seems to be no clear evidence that volume ratios have any effect on the adherence of films either at oxidation temperatures or at room temperature after high temperature oxidation.

The pre-existing film and phase changes

During the heating-up of a metal to high temperatures for isothermal oxidation, a film will be formed which will undergo changes in composition with temperature, and changes in stress due to expansion of the metal relative to the film. Normally the coefficients of thermal expansion of the metals are somewhat greater than those of the oxide, and the latter would therefore tend to crack. However, this will be to an unknown extent offset by the large volume ratio of the oxide.

The effect of the pre-existing film was investigated by Païdassi[12] on iron over the range 400–700 C, and it was found that its presence resulted in a considerable increase in the weight-gain over the range 400–600 C.

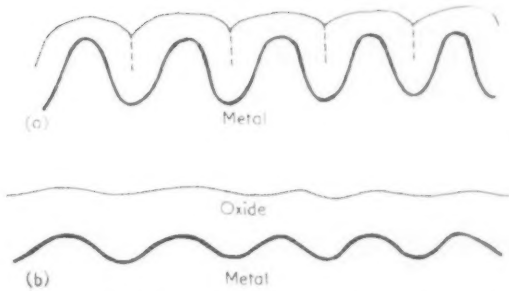
Phase changes will also cause a change in the dimensions of the metallic substrate which will impose temporary stresses, usually compressive, on the growing film. Any effects due to pre-existing films or phase changes can be avoided in some cases by heating-up in suitable reducing atmospheres, but in the majority of cases the oxide films are not reducible and these factors have to be considered.

Surface roughness

Rhodin[13] concluded that small irregularities on copper increased the surface area by a factor of 2·45, but that these fissures or cracks had such small dimensions that they were obliterated by the growing oxide film during the first few minutes of oxidation at low temperatures (i.e. by films 100 Å thick). These irregularities could be removed completely by electrolytic polishing.

Païdassi, in his work on iron, assumed that electropolishing was a prerequisite for adherence, and the fact that his films are more adherent than others on iron of similar purity and formed over the same temperature range, suggests that his view is justified.[12, 14]

If surface roughness is having an effect on adherence, it is probably due to oxide filling cracks in the surface and becoming highly stressed due to the effect of its larger volume. Although this effect will tend to disappear as the film thickens, due to the smoothing effect noticed by Rhodin, it will leave, at low temperatures at any rate, a highly stressed zone close to the metal surface (*see* Fig.1*a*). The compressive stresses present in the cracks would not necessarily lead to exfoliation as they would exert a 'keying-on' effect.



1 *Effect of surface roughness on oxide film. (a) first stage, (b) second stage*

Assuming growth by outward metal movement, the next stage would be the growth of the film in the vicinity of the asperities and their gradual disappearance (Fig.1*b*).

Evans[15] has shown that internal stresses are less marked in films on finely abraded metal than on coarsely abraded, which is in keeping with the effect of roughness suggested above.

The effect of film thickness

Evans[15] has shown that thin oxide films are highly stressed in compression on the surface attached to the metal. The stress gradient decreases as the thickness increases suggesting that the upper layers are relatively stress-free. It is not difficult to see how this stress gradient can arise, since in the early stages of oxidation the film forms by the enlargement of nuclei of oxide within the metal. Later growth could be stress-free, by the outward movement of metal through the film, forming oxide on a free surface.

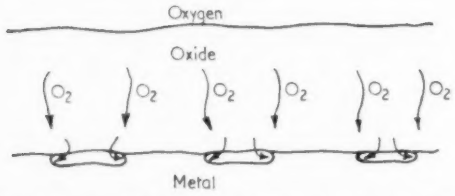
Many films, such as those of Cu_2O on copper at high temperatures, become more adherent as they grow thicker. Such films are comparatively unstressed due to their inherent plasticity at high temperatures.

In the absence of free surfaces such as edges, it is not possible for a film to inherit a stress normal to the interface on cooling. The only stress present would be a shear stress on the metal-oxide interface.

In the majority of cases the film-metal interface will derive a normal stress from the presence of edges or discontinuities. Thick films, due to their inherent stability, will not easily exfoliate. However, in thin films this stability does not exist and films which possess compression stresses due to differential contraction on cooling are likely to become non-adherent. In the case of thick oxide films on thin metal the film would be strong enough to impose tensile stresses on the metal during the cooling process which the metal could dissipate by deformation.

Due to the formation of holes by the coalescence of vacancies in films growing by outward metal diffusion, stresses at the interface may be continually reduced as the film thickens. But any such process will be offset by any inward diffusion of oxygen.

The adherence of oxidation-resistant films on high temperature alloys seems to be the result of a different mechanism, since adherence seems to decrease with increasing thickness. Here there are so many competing factors that it is difficult to put forward a simple explanation. However, these alloys seem to possess



2 *Hole formation and subsequent filling by oxidation of remaining metal conduits*

high strength films in conjunction with subscale. Since subscales are the result of oxygen diffusion, they may be the seat of considerable stress.

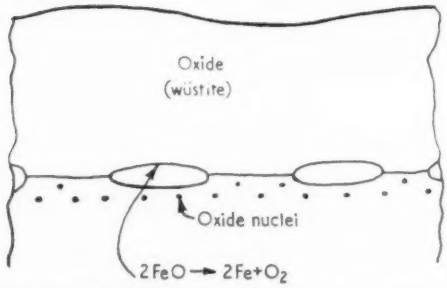
Dominant diffusion mechanism

In the early stages of oxidation, coherent films are formed as a result of the inward diffusion of oxygen and the growth of the oxide nuclei so formed. Further growth, in many cases, is maintained by outward diffusion of metal ions. It would be expected that, in these cases, internal stresses in the vital interfacial area would be less than if the film continued to grow by inward diffusion of oxygen. However, the outermost layer of metal becomes a metal consumption zone, and if growth was wholly by outward diffusion presumably it would be possible for the film to lose contact with the metallic substrate. In films on copper and iron this tendency does arise and holes have been formed in the centre of wires. It would be possible also for voids to be formed by internal stresses giving rise to blisters, and not necessarily as a result of metal consumption.

Dunnington *et al.*[16], on the other hand, believe that the vacancies diffuse into the metal causing the formation of holes on the oxide-metal interface to be postponed until the metal itself is saturated. At some stage the metal and oxide part company on one interface, the resulting hole acting as a sump for the vacancies present in the metal and those arriving at the other interface. In this way contact can be maintained on one metal-oxide interface until all the metal is consumed. There is no doubt that some such effect is operative at high temperatures (>1000°C) on iron, but this alone is not sufficient to explain the adherence and lack of visible holes on some metal-oxide interfaces at lower temperatures.

Dravnieks and McDonald[17] consider the formation of holes in the base of the film due to outward ionic movement, and postulate either the collapse of the film into these holes or the filling of them with new oxide. The latter process would require the diffusion of oxygen inwards or dissociation of the scale at the surface of the hole causing release of oxygen.

A consideration of this mechanism shows that as soon as large cavities are formed the metal supply to the scale is reduced. Dravnieks supposed that thickening and oxidation of the remaining metal conduits, by diffusion of oxygen inwards, was the most likely so that oxide would form and fill the cavities (Fig.2). Lattice diffusion of oxygen inwards seems to be very unlikely in the majority of oxides, but would be possible in the case of CuO, ZrO_2, and TiO_2, i.e. oxides that are believed to have vacancies in the oxygen lattice. On the other hand, diffusion along the grain boundaries would be possible in all cases.



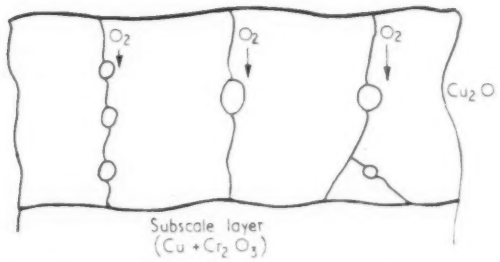
3 *Oxide dissociation and subscale formation below porous layer*

Most of the work on the oxidation of steels (e.g. Pfeil[18] and Sachs[19]) shows that there is an extremely porous scale layer at the base of the scale, the scale tending to be less porous as one moves towards the oxygen interface. The absence of porosity in the outer layers is due partly to the fact that they lie outside the metal consumption zone and partly that they can permit oxygen diffusion inwards. There seems no doubt that the Dravnieks mechanism causes porosity in the wüstite layer, and probably the main reason why it is not seen in copper at the same temperature is because of the lower strength and the high plasticity of Cu_2O.

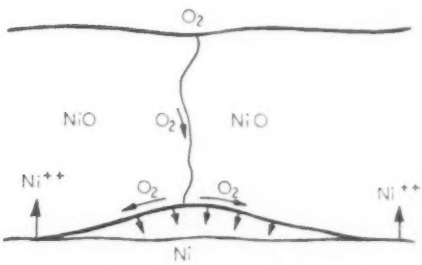
The fact that internal oxidation continues to take place after films of the metal deficit type such as Cu_2O and wüstite have been formed, shows that oxygen is getting through the scale and through the porous layer. The mechanism for getting oxygen through the scale is obscure; however, in cases where porosity exists in it, it would be possible for oxide to dissociate (according to Dravnieks) at the oxide/metal surface, and at low pressure dissolve in the metal below it and form internal oxide (Fig.3). To do this it must first pass through the film by: (*a*) oxygen vacancy mechanism (only possible in some oxides), (*b*) rifts (possible in the non-recrystallized zone), (*c*) grain boundaries, and (*d*) holes communicating with the outside surface. Regarding (*d*), photomicrographs of oxide films on copper-chromium alloys show holes in the Cu_2O along the grain boundaries (*see* Fig.4, after Dennison[20]). Since grain boundaries are zones of misfit it would be comparatively easy for the large oxygen atoms to filter along them, until they reach a hole where the process could be speeded up by oxide dissociation.

According to Czerski and Franik[21] the oxidation of 99·8% purity nickel at 1300°C proceeds mainly by inward oxygen penetration along the grain boundaries, and only 30% of the thickness is due to metal movement outwards. A similar result was obtained for nickel by Ilschner and Pfeiffer[22] at 1000°C. Czerski and Franik also claim that the cracks formed by the increase in volume $\left(\frac{NiO}{Ni}=1{\cdot}6\right)$ are filled by oxide produced by the incoming oxygen in the oxide-metal interface, and that it is the degree of compensation occurring by this process which decides the adherence (*see* Fig.5). Arkharov *et al.*[23] are also of the opinion that oxygen diffuses inwards towards the NiO/Ni interface.

Where the grain boundaries meet the oxide/metal interface, points of high stress are present due to local oxidation of the nickel in this region and the consequent increase in volume, even when allowance has been made for solution of oxygen in the adjacent metal.



4 *Oxidation of a 1·2%Cr–Cu alloy at 950°C (after Dennison[20])*



5 *Oxidation of Ni according to Czerski and Franik[21]*

Forces are developed which tear the film from the metal at the interface on both sides of the grain boundary and facilitate the entry of oxygen to the metal surface, which allows the re-establishment of contact between the oxide and the metal.

In the case of copper the tendency to form voids is overcome, at least at temperatures in the vicinity of 1000°C, by the plasticity of the film which allows it to shrink-in on the consumed metal. Moore[24] believes that this plastic flow is caused by excess of free energy due to the coalescence of vacancies at the metal/oxide interface. The forces must be very high since a nichrome wire wound round a copper specimen was found to have its length reduced by 3%. In the case of oxidized phosphorus-containing copper wires, a marked segregation of liquid or viscous copper phosphate had taken place in the centre of the oxide wire formed.[25]

Uhlig,[26] using iron wire containing 2%Al and 2%Si, found a large hole in the middle after oxidizing at 1165°C. The existence of this hole can probably be explained by the formation of an inner rim of SiO_2 and Al_2O_3 preventing inward plastic movement of the iron-rich outer scale. It would be expected that a thick film on such an alloy would have poor adhesion.

Engell and Wever's[27] results on iron agree with other results[21, 22] on nickel in that they show that inward and outward diffusion seem to be equally important at high temperatures leaving platinum markers embedded in the oxide near the original interface position. However, at temperatures below 800°C the markers which Engell and Wever used were maintained in position on the interface, probably by plastic oxide flow.

The conclusion is that there should be no essential difference as regards exfoliation between inward and outward growing oxides. In the case where outward metal diffusion is known to predominate, inward oxygen diffusion can fill any holes forming at the interface. Where inward oxygen diffusion predominates stresses can be relieved by deformation or blistering within the film.

Relative coefficients of thermal expansion

Exfoliation can either occur upon oxidation or during cooling after oxidation. The former process can be detected on iron because it often gives rise to different proportions of the three oxides in different portions of the same specimen. When the film breaks away from the specimen during oxidation the supply of metal ions to the film is interrupted, which increases the proportion of higher oxides in the film. This is very clearly shown by the conflicting results for the composition of oxide films on iron at various temperatures. Païdassi[28] shows that at temperatures from 700 to 1000°C the film on pure iron is 95% wüstite. Bénard and Coquelle[29] obtained large amounts of Fe_3O_4 and Fe_2O_3 at the higher temperatures due to exfoliation of the film on oxidation, as did Engell and Wever[27] for iron wires.

Many films, which in the case of iron contain very high proportions of wüstite, show the presence of cracks between the film and the metal. These cracks have clearly been produced during cooling, and presumably arise from differences in the coefficient of thermal expansion of film and metal.

The fact that the contraction of the metal will in general be greater than that of the oxide would be expected to impose a compressive stress on the latter which could lead to exfoliation on cooling, or possibly during oxidation due to fluctuations of temperature.

In general the coefficients of expansion of oxides at room temperature are only a little less than the metals from which they are formed, except in cases like cuprous oxide where the difference is an order of magnitude (Table II).

Since the oxide film formed on pure copper at 900°C is extremely adherent and only comes off after considerable bending, it would seem that differences in expansion coefficients need not be important by themselves. The small periodic fluctuations that take

TABLE II Expansion coefficients

Metal/oxide system	Oxide		Metal		Reference
	Coefficient, $\times 10^{-6}$	Temperature range, °C	Coefficient, $\times 10^{-6}$	Temperature range, °C	
Fe/FeO	9·4	20	11·7	20	Rigby[30]
	12·2	100–1000	15·0	0–700	Cleaves and Hiegel[31]
Co/Co_3O_4	13·6	20–900	12·4	40	Metals Handbook[32]
			18·0	25–350	Lucas[33]
Cr/Cr_2O_3	7·3	100–1000	8·2	20	Lucas[33]
			9·5	0–1000	Rigby[30]
Cu/Cu_2O	0·94	20	16·7	20	I.C.T.[34]
			20·0	0–1000	Metals Handbook[32]

place during isothermal oxidation using some form of on-off control may have an effect in the case of non-ductile oxides. In the case of thick scales it is probable that such variations do not penetrate to the oxide/metal interface.

High temperature plasticity of scales

As far as adherence at elevated temperatures is concerned the plasticity of the scale has a considerable effect. The tensile strength and ductility of cuprous oxide at elevated temperatures has already been measured by the author,[25] and the ductility has been shown to be quite considerable when the copper is pure, but markedly reduced by small amounts of certain impurities. These impurities not only embrittle the film leading to its exfoliation on cooling, but also cause a reduction in the adherence of the film when the metal is worked at high temperatures.

Mackenzie and Birchenall[35] have recently shown that wüstite has appreciable plasticity enabling it to creep under stress at 900°C. They also showed that scales formed in H_2–H_2O mixtures, consisting only of wüstite, were adherent; whereas scales formed in oxygen which therefore contained small amounts of Fe_2O_3 and Fe_3O_4, were not adherent. This they explained by the inhibiting effect of the more creep resistant oxides Fe_2O_3 and Fe_3O_4 on the plasticity of wüstite, preventing the latter from following up, and adhering to, the shrinking metal surface. However, it is possible that the low partial oxygen pressure itself is responsible for the improved adherence since there will be a reduction in the relative amount of oxygen diffusing inwards through the film compared with metal diffusing out. The results will be to reduce the stress at the oxide/metal interface due to the increase in volume on the oxidation of metal in this area.

Metal surface configuration and keying-on of oxide

Examination of photomicrographs of oxide films on metals shows marked differences in the nature of the oxide/metal interface. In some cases this surface still remains smooth after the build-up of a considerable thickness of oxide. In others the film is very uneven and shows in some cases a strong tendency to penetrate the grain boundaries of the metal. It would seem that this penetration might influence the adherence of the film.

The life of a nickel–chromium alloy subjected to intermittent oxidation tests at high temperatures is influenced by small amounts of silicon and alkaline earths. It is claimed by Lustman[36] and others that exfoliation is reduced by the restriction of outward diffusion of nickel by the presence of trace elements in the oxide lattice. This favours the inward diffusion of oxygen which causes more marked internal oxidation, and a rougher metal/oxide interface which in turn favours the keying-on of oxide. It was also noticed that the more resistant the films were to spalling, the higher the Cr_2O_3 content of the film and the lower the amounts of NiO and $NiO.Cr_2O_3$. The latter oxides are only formed by the outward diffusion of Ni through the Cr_2O_3, which process is effectively blocked by the presence of small Si^{++++} ions. On the other hand there appears to be some evidence that the presence of Si and alkaline earths also reduces the rate of continuous oxidation, in which case the reduction of spalling may merely be due to the existence of a thinner film.

Gulbransen and Andrew[37] claim that a good crystal structure match between oxide and metal is necessary for maximum adherence. Photomicrographs of adherent oxide films on metals often show clearly that recrystallization has occurred and there is no structural continuity between metal and oxide. Only at low temperatures or at low oxygen partial pressures is there any evidence of such structural continuity.

Roughening of the interface can occur during oxidation due to tensile stresses in the metal brought about by films of high volume ratio, such as was found in zirconium by Hérenguel *et al.*[38] While such films are fissured in directions both normal and parallel to the interface, their general adherence is good. In this particular case adherence was probably assisted by a considerable elongation of the metal itself, which has been shown to occur to an extraordinary extent by Phalnikar and Baldwin.[39]

Wagner[40] has shown, in a theoretical analysis, how the formation of rough oxide/alloy interfaces may be explained in the case of alloys containing noble metals. Such interfaces are likely to form in all cases where diffusion within the alloy is considerably less rapid than the diffusion of the less noble metal within the oxide. In such cases the interface consists of inward protruding sections of the oxide of the less noble metal interspersed with slender trunks of alloy rich in the more noble metal. Although the degree of adherence has not been reported in many of the cases of this type examined, there is little doubt that the interpenetration of oxide and metal in this way leads to an improvement in the adherence.

Effect of purity

As far as the adherence of films on pure metals is concerned, one of the most important factors is the degree of purity. Small amounts of preferentially oxidizable impurities appear to accumulate at the oxide/metal interface and tend to increase the stresses at this point. Whereas films on electrolytic copper and nickel are normally adherent after high temperature oxidation, films on iron are not, unless the iron is of the highest purity, i.e. spectrochemically pure. How this arises is clear from recent work by Païdassi[41] and Moreau[42] in which the accumulation of the oxides of impurities on the interface is shown microscopically.

It is well known that the lower layers of oxide films tend to have a fine grain-size while the upper layers are coarser. This seems to be due to the accumulation of easily oxidizable impurities inhibiting grain growth. In the outer layers of the film, where the purity is greater, large grains form, although in some cases intergranular films rich in impurity elements can be seen to be present in these layers.[25]

The fact that 0·04% of phosphorus can form a phosphate film leading to high temperature weakening and exfoliation of copper scales on cooling, shows how certain impurities can be effective. Small amounts of silicon are preferentially oxidized in steels and are present in the grain boundaries of the wüstite as fayalite, which forms a low melting point eutectic with wüstite, melting at 1170°C.

The effect of oxidizing atmosphere

Since adherence is probably the result of a delicate balance between hole formation and oxide formation on the metal/oxide interface, reduction of the oxygen

partial pressure might be expected to have an initial beneficial effect, although sooner or later the formation of holes might weaken the interface mechanically.

Wüstite films formed on iron in steam–hydrogen mixtures $\left(\frac{H_2O}{H_2}=1{\cdot}5\right)$ were found by Chaudron and Collongues[43] to be adherent in conditions that otherwise do not produce adherent films upon iron.

Armco iron, containing 0·037%C and 0·043%Mn among other elements, gave rise to an irregular oxide/metal interface containing small fissures and holes. In no case, however, did the film cease to be adherent, whereas in air, films on Armco iron would not have been adherent.

Reduction of the partial oxygen pressure in air by dilution with nitrogen appears to have very little effect on adherence, although it does appear to modify the properties of the oxide film on commercial steels.[44]

The fact that steam–H_2 mixtures produce adherent films of wüstite on iron may be due to dissociation and the individual effects of hydrogen and oxygen. Both gases can diffuse through the film, but the small amount of oxygen now available will quickly combine with the outward diffusing iron and be incorporated as wüstite on the outer surface. When hydrogen is present in excess it reduces the wüstite on the outer surface, but Richardson and Dancy[45] found that the resulting iron then diffuses back to the oxide/metal interface. During oxidation, with rather greater oxygen pressures, it is possible that the hydrogen which would readily diffuse into the bulk of the film, reduces the oxide and so decreases the stresses within it. The resulting iron could then diffuse back to the oxide/metal interface, filling holes left by any outward diffusing iron. More probably it would diffuse to the outer surface providing some of the iron needed for film growth and so slow down the development of a porous zone on the metal/oxide interface. Once the growth of this zone has been reduced any plasticity present in the film would be more likely to make itself felt by enabling oxide to maintain contact with the metal by a creep process.

How far the adherence of films formed on iron in H_2–H_2O mixtures is due to the non-existence of the stronger and less plastic oxides such as Fe_3O_4 and Fe_2O_3 has yet to be determined.

CONCLUSIONS

This brief consideration of the factors that could affect adherence of oxide films on metals shows that film plasticity and relative expansion coefficients could be important. As far as pure metals are concerned the effects of minor impurities and the oxidizing atmosphere are very great.

The diffusion mechanism of oxygen ions inwards or metal ions outwards definitely plays an important part, and it is considered that, in the main, good adherence is the result of a balance between the two.

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The physical metallurgy of 12% chromium steels

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SYNOPSIS

The paper details the effects of alloying elements on the phase equilibrium, the critical ranges, and the martensite temperature ranges, in low-carbon 12% chromium steels. The effect of alloying elements on the tempering characteristics and especially the secondary hardening characteristics of 12%Cr has been considered in detail, with special reference to the effect of the alloying additions on the formation of the M_2X precipitate, The mechanical properties of alloyed 12%Cr steels have also been investigated and it has been shown that secondary hardening is accompanied by a decreased proof ratio and poor impact properties. Additions of up to 2%Mo have been shown to improve the impact properties, and the effect of V has also been considered. Above 0·3%V the impact resistance has been shown to decrease and for the best combination of mechanical properties a 0·1%C, 12%Cr, 2%Ni, 1½%Mo, 0·3%V steel is preferred, but where better impact and ductility are required the V may be omitted. There are two temperature ranges in which it is advantageous to temper alloyed 12%Cr steels, namely 250–300°C and 600–700°C. In these tempering temperature ranges attractive mechanical properties can be obtained, but tempering in the range 450–500°C has been shown to be most undesirable. 1813

INTRODUCTION

STAINLESS STEELS which are commonly used fall into several main types with different combinations of corrosion resistance and mechanical properties. One important type is the 12%Cr transformable steel which is finding increasing applications as a high strength stainless structural steel. In a systematic programme of work the structure and properties of a large number of 12%Cr steels have been studied so that the effect of alloying elements is known and it is therefore possible to develop steel compositions for specific applications.

It is helpful to compare the 12%Cr steel with the other types of stainless steels because the critical importance of composition can then be appreciated. These different types result from the effect of chromium on the equilibrium diagram for the Fe–Cr–C system. One important effect of chromium is, of course, that it controls the degree of corrosion resistance. Equally important, however, is the effect on the equilibrium diagram because this controls the structure obtained and hence the mechanical properties.

An indication of the structures obtainable in these steels can be seen from Figs.1*a* and *b*.[1] In Fig.1*a* the effect of variation in chromium content is shown with a constant carbon content of 0·1%. The most important feature of this diagram is the austenite phase field since it is only steels which are austenitic at the solution temperature which transform on cooling. It will be seen that the temperature range of austenite stability decreases with increase in chromium content until with more than 13–14%Cr it is difficult to ensure a fully transformable steel. Any ferrite which is present at the solution temperature remains in the structure to reduce the strength and also to detract from other properties. Consequently, at a low carbon content of about 0·1% (which is desirable for ease of welding and good ductility) there is a choice between the following compositions:

(i) a low-chromium steel (i.e. 1–5%Cr) which has insufficient chromium to provide much increase in oxidation resistance, but is transformable with increasing hardenability as the chromium content increases

(ii) an 11–13%Cr steel which is fully transformable and air hardening up to large section sizes. The oxidation resistance is adequate at temperatures up to 700°C and there is a possibility of adding alloying elements to improve the tempering resistance without introducing large amounts of ferrite

(iii) a steel containing 17% or more of chromium which will be non-transformable, i.e. fully ferritic. Although the oxidation resistance is improved, there is now no possibility of obtaining high tensile properties, except by cold working or by the utilization of a precipitation hardening effect.

Of these alternatives there is greatest scope for development in the 11–13%Cr steels, which is the reason for the programme of research work which will be described.

One other aspect which must be considered is the effect of alloying elements in depressing the martensite transformation range. A 0·1%C low-alloy steel has an

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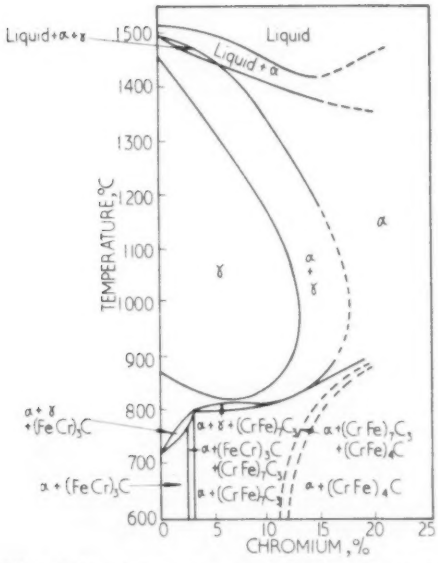
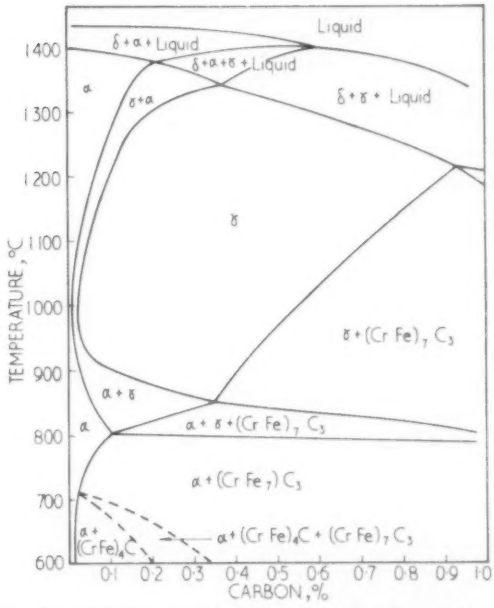


1a *Effect of chromium content on equilibrium relationships in 0·1%C steels*



1b *Effect of carbon content on equilibrium relationships in 12%Cr steels*

M_s temperature of about 450°C which means that complete transformation to martensite can occur when cooling from the solution temperature to room temperature. There is a decrease in this M_s temperature when increased amounts of alloying element are added[2] and the M_s temperature of a 0·1%C–12%Cr steel is about 300°C. Since the martensite transformation temperature range is about 150°C this still allows complete transformation to martensite on cooling to room temperature. On the other hand, it can be seen that the scope for adding alloying elements which will depress the martensite transformation range still further is limited, otherwise the M_f is depressed below room temperature and incomplete transformation is obtained. As an indication of this effect, the addition of 3%Ni to a 0·1%C–12%Cr steel depresses the M_f to approximately 50°C and thus represents about the maximum addition of nickel that can be made.

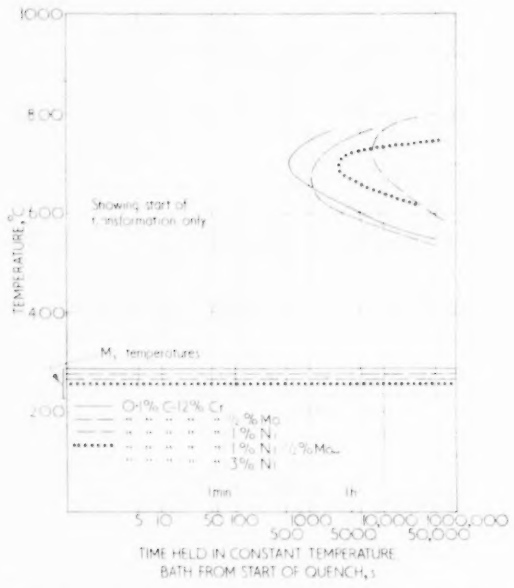
It will be appreciated that providing the martensite transformation temperature range is known, transformation can be obtained by using an appropriate transformation treatment, such as refrigeration. These steels can then be classified as controlled transformation steels and a full description of this type of steel has been given recently.[3] On the other hand, if the martensite transformation temperature range is so low that it is impossible to produce transformation then these steels are stable austenitic steels and can be used as such.

These stainless steels, the high chromium fully ferritic, the controlled transformation, and the stable austenitic, each have commercial applications. It was felt, however, that the 12%Cr type had not achieved its full potential because the possible alloy combinations were not fully understood. There were considerable difficulties in developing suitable steels because of the critical nature of the composition. A minimum of 11%Cr is necessary to ensure adequate oxidation resistance. To obtain reasonable ductility and impact properties and to prevent welding difficulties a low C content is desirable. The combination of 0·1%C and 12%Cr produces a steel near to the borderline of the stable austenite phase at solution temperatures and with an M_f temperature at about 100–150°C. The mechanical properties of the base steel can be improved by the addition of alloying elements such as molybdenum and vanadium. Since these are strongly ferrite-forming they must be balanced by austenite-forming elements such as nickel, manganese, copper, and cobalt. This combined alloy addition may then depress the martensite transformation temperature range so that there is incomplete transformation and the mechanical properties are impaired.

It will be seen, therefore, that there is only a limited possibility for alloying 12%Cr steels and the object of the work which is described in this paper was to define these limits.

THE EFFECT OF ALLOYING ELEMENTS ON STRUCTURE

The most important aspect of structure is for it to be free from ferrite at the solution temperature. A solution temperature of 1050°C is adequate unless very strong carbide-forming elements are used, and consequently it is essential to have a ferrite-free structure at this temperature. If welding is to be used, however, parts of the heat-affected zone will be heated to a higher temperature than this so that it will then be necessary to study the structure over a wide range of solution temperatures. The diagrams in Fig.1 show that the austenite phase field contracts with increasing temperature. Consequently, any alloy selected for commercial development should be well within the austenite phase field at 1050°C in order to ensure a ferrite-free structure at, say, 1300°C. Generally speaking, however, most of the preliminary investigation has concentrated on the structures at 1050°C since it is a simple matter to modify any selected composition to achieve a stable structure at higher temperatures.


1000
800
600
400
200
TEMPERATURE, °C
Showing start of transformation only
M_s temperatures
0·1% C-12% Cr
" " " " + 1% Mo
" " " " + 1% Ni
" " " " + 1% Ni + 1% Mo
" " " " + 3% Ni
1min
1h
5 10 50 100 500 1000 5000 10,000 50,000 100,000
TIME HELD IN CONSTANT TEMPERATURE BATH FROM START OF QUENCH, s

2 *Isothermal transformation diagrams for various* 12%Cr *steels*

It was necessary to fix a base composition for this work and this was chosen as 0·1%C, 12%Cr, 0·5%Mn, 0·25%Si, 0·02%N. Because of the affinity of high-chromium steels for nitrogen and the presence of nitrogen in ferrochromium it is difficult to obtain a lower nitrogen content without a special steelmaking practice, and consequently 0·02%N represents the normal nitrogen content of this type of steel without any deliberate alloy additions. The steels used for this work were made as 25 or 112 lb air-melted induction furnace casts and were then rolled or forged to $\frac{3}{4}$-in. dia. bar. This bar material was then heat-treated before metallographic examination and mechanical testing.

If the microstructure of this steel is examined after solution treatment it will be found that there is no ferrite after heating for 1 h at temperatures between 1050° and 1250°C, there is about 50% ferrite after heating at 1300°C and the structure is completely ferritic after heating at 1350°C. These findings agree well with the phase boundaries shown in Fig.1. They indicate that the base composition is as near to the ferrite phase boundary at temperatures between 1050° and 1300°C as allowable and consequently any addition of a ferrite-forming alloying element must be balanced by an equivalent amount of an austenite-forming element. With a base composition near to the phase boundary in this way it is possible to summarize the effect of alloying elements by determining the increase or decrease in ferrite content produced by a unit increase in alloy element. The effect is not always linear but a linear extrapolation is sufficiently accurate for purposes of alloy development. To determine the effect of ferrite-forming elements, small additions were made to the base composition since this was already a borderline composition. To determine the effect of austenite-forming elements a base composition which has about 20% ferrite at 1050°C is necessary and the same composition as the base steel but with 15%Cr is suitable. The summary of this part of the work is as follows:

Element	Change in delta ferrite, % for 0·1 wt-%	for 1 wt-%
N	−22	
C	−21	
Ni		−20
Co		−7
Cu		−7
Mn		−6
Si		+6
Mo		+5
Cr		+14
V		+18
Al		+54

Elements such as Ti and Nb are difficult to assess in this way because in addition to increasing the ferrite content by their effect on the phase boundary, they also remove carbon from solution, which has the effect of increasing the ferrite content still further. This effect may occur to a lesser extent with vanadium and also with Al (owing to a combination with nitrogen) so that the values quoted for these elements are not very accurate, but at least they show that there is a strong effect. Very approximate assessments of the ferrite-forming tendencies of Nb and Ti give an increase of 12% and 14% respectively per 1% of element added.

A similar study has previously been carried out on the effect of alloying elements on a 17%Cr–4%Ni base composition.[3] Reasonable agreement was obtained between these two series of results, which indicates that they are sufficiently accurate to be used for alloy development work. The agreement was best with the alloying elements that have strong effects and this is to be expected since some variation is always obtained with borderline compositions.

THE EFFECT OF ALLOYING ELEMENTS ON ISOTHERMAL TRANSFORMATION

The addition of 12%Cr to a low-carbon steel has a marked effect in retarding the ferrite and pearlite transformations and also the bainite reaction so that the steel is air hardening. There is interest in the isothermal transformation diagrams of alloyed 12%Cr steels to indicate the degree of martensitic hardenability and the possibility of obtaining pearlitic and bainitic transformations. Some typical diagrams are shown in Fig.2 from which it is seen that the rate of transformation is extremely slow even in the base composition. Nickel is particularly effective in retarding transformation and when more than 1% is present it is unlikely that any appreciable amount of transformation will occur even during prolonged heat-treatment at 700°C. The commercial aspect of this is that while there is little difficulty in obtaining complete transformation to martensite even with air cooling in large section sizes there is likely to be considerable difficulty in softening for machining or fabrication.

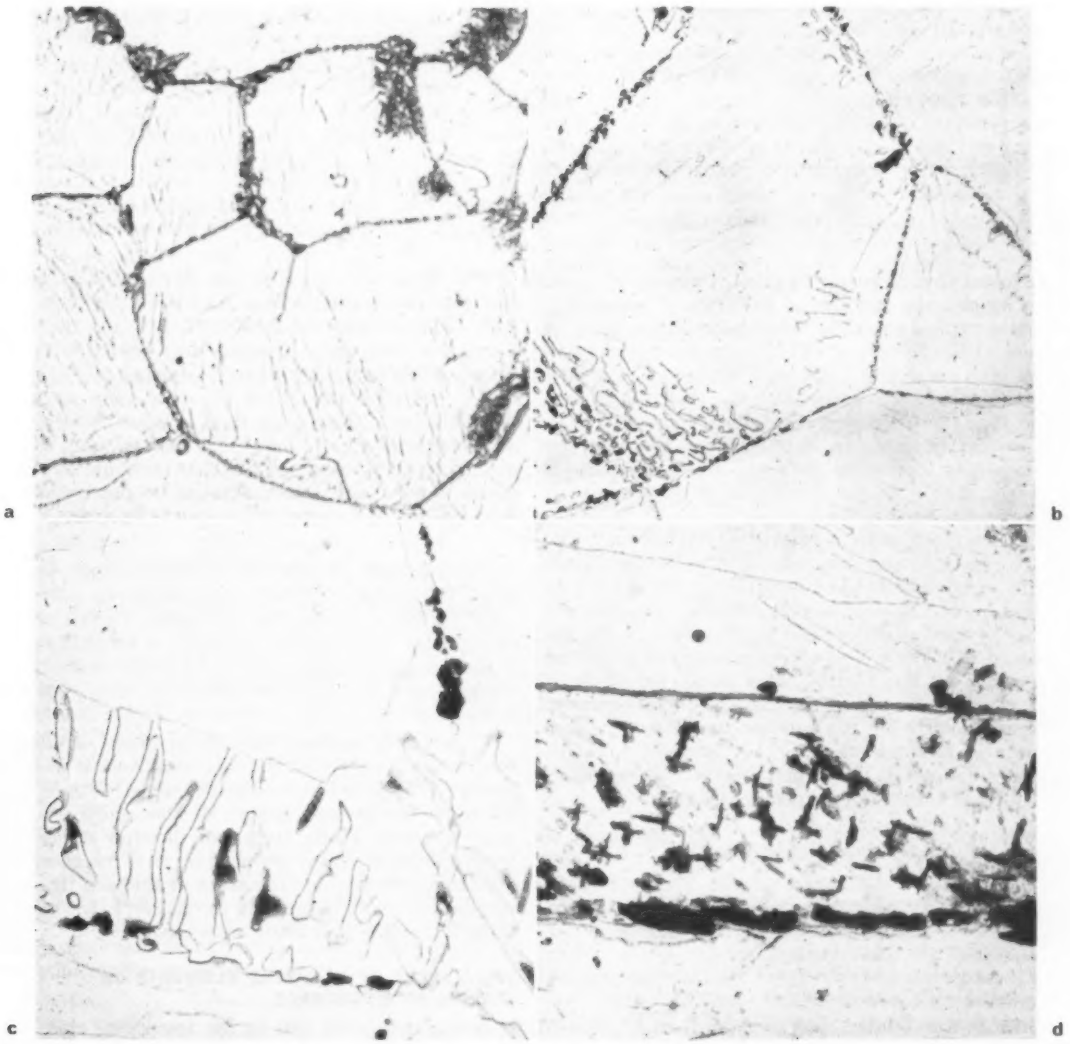
Another interesting feature of the transformation diagrams of this type of steel is that there is a tendency for carbide precipitation to occur before pearlite formation. Figure 3 shows this effect in a 12%Cr–Ni–Mo–V steel. The optical micrograph (*a*) indicates a small amount of transformation after 65 h at 650°C but the electron micrographs show that a grain boundary network of carbide has precipitated before the formation of some very coarse pearlite (*b* and *c*).

Since this precipitate could have an adverse effect on ductility it indicates that prolonged heat-treatments of this type (or extremely slow furnace cooling) should be avoided. There would appear to be no possibility of obtaining bainitic transformations in these steels. The reduction of chromium locally due to the precipitation of grain-boundary carbide, raises the M_s and also gives plates of autotempered martensite adjacent to the grain-boundary carbide (*d*).

THE EFFECT OF ALLOYING ELEMENTS ON TRANSFORMATION TEMPERATURES

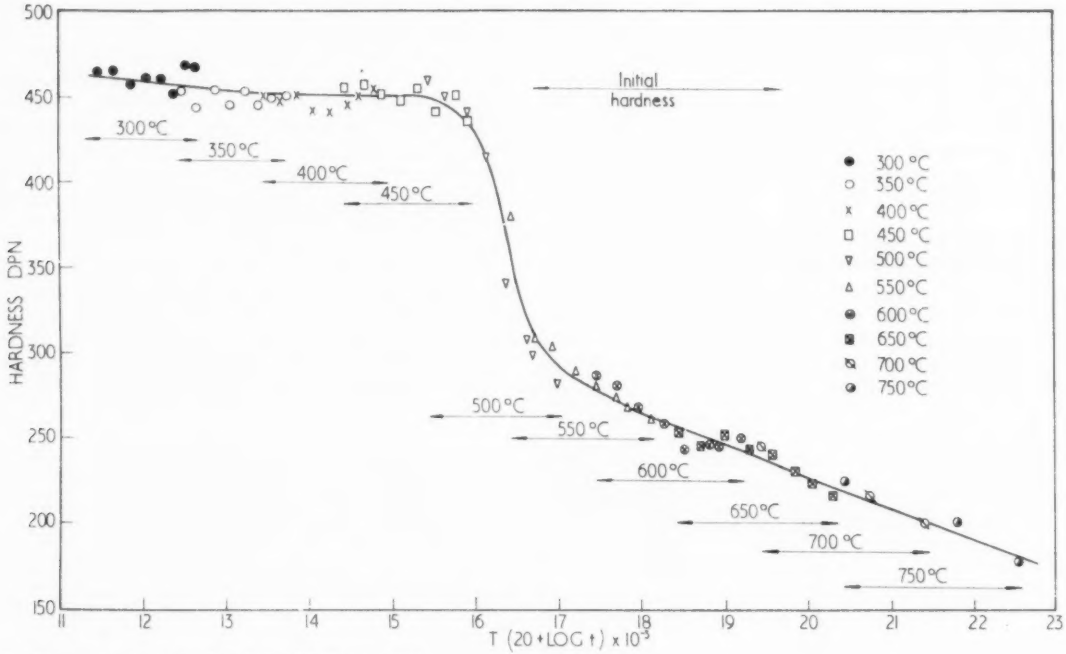
There is particular interest in the effect of alloying elements on the A_{c1} and M_s temperatures. The A_{c1} temperature has a direct bearing on tempering treatments. It will be seen later that tempering in the range 600–700°C is necessary in many 12%Cr steels to obtain the best combination of mechanical properties. Furthermore, prolonged tempering at 700–725°C is the best way in which the steel can be softened for machining or fabrication. The A_{c1} temperature of the base composition is about 740°C and some alloy additions depress this. Since re-austenitization is undesirable during tempering it is necessary to know the effect of different alloying elements on the A_{c1} in order that a limit can be set to the tempering temperature for any particular steel. Preferably the A_{c1} temperature should not be less than 700°C.

As it is the effect of the A_{c1} on the response to tempering which is of great importance, dilatometrically determined A_{c1} points give little information about the probability of re-austenitization during tempering because of the greater approach to the equilibrium condition during isothermal tempering. We



a optical micrograph ×1500 *b* electron micrograph showing grain-boundary carbides and pearlite ×3000 *c* as *b* ×15000
d electron micrograph of autotempered martensite adjacent to grain-boundary carbide ×15000

3 *Microstructures of isothermally transformed specimen of* 0·1%C, 12%Cr, 2%Ni, 1½%Mo–V *steel transformed 65 h at 650°C*



4a *Tempering curve for 0·14%C–12%Cr base steel*

have found that re-austenitization occurs at temperatures which are a function of the time of tempering, and in a typical example, whereas austenite may be formed within 20 min at 750°C, it may not be formed in less than 5 h at 700°C or lower. It therefore becomes very difficult to determine the A_{c1} and it has been found that the tempering curve itself is probably the best indication of the re-austenitization temperature, the hardness increasing suddenly when austenite is reformed, owing to the formation of martensite on subsequent cooling.

The effects which have been observed can be stated briefly. Carbon and nitrogen within the relatively narrow range which has been studied appear not to depress A_{c1} but Mn and Ni very definitely do. Nickel is by far the most effective, and 3%Ni lowers A_{c1} to the region of 650–700°C and a similar effect is observed with about 4%Mn. On the other hand, vanadium is most effective in raising the A_{c1} and molybdenum behaves similarly but is less effective. Aluminium and silicon also tend to raise the A_{c1} while high cobalt additions depress it. Very approximate assessments of the effects of these alloying elements follow:

	Depression of A_{c1} by 1% alloying element, °C
Nickel	−30
Manganese	−25
Cobalt	−5
Silicon	+20–30
Aluminium	+30
Molybdenum	+25
Vanadium	+50

It can be appreciated that no more than 3%Ni can be accommodated if the steels have to be tempered at 650°C and if re-austenitization must not occur below 700°C, 2%Ni is the limit and preferably should be combined with both molybdenum and vanadium.

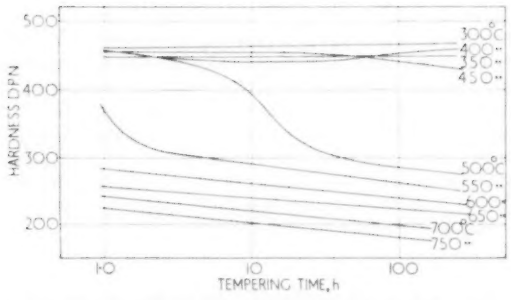
The M_s temperature is also important because it controls the final structure obtained. It should desirably be above 200°C in view of the fact that the transformation temperature range is at least 150°C. The effect of alloying elements in depressing the M_s below 300°C, which is the M_s of the base composition, is shown below. Most of the data is taken from Steven and Haynes[2], and although not derived from 12%Cr steels appears to agree with our own observations. These results were supplemented by determinations for tungsten and silicon by Payson and Savage.[4]

	Depression of M_s by 1% alloying element, °C
Carbon	−474
Manganese	−33
Nickel	−17
Chromium	−17
Molybdenum	−21
Tungsten	−11
Silicon	−11

It should be stressed that these effects of alloying elements will vary with the base composition used and should be regarded as indicative only. However, they do show the critical nature of the composition of 12%Cr steels. While trying to develop steels with improved mechanical properties, it is necessary to maintain a ferrite-free structure at solution temperature and to avoid depressing the A_{c1} and M_s temperatures below 700° and 200°C respectively.

THE EFFECT OF ALLOYING ELEMENTS ON TEMPERING RESISTANCE

It is useful to refer first to the tempering characteristics of the 12%Cr base composition. It is convenient to use a tempering curve based on a combined time/temperature parameter T(20+log t) where T is the temperature °A and t is the time in hours. In this way



4b *Isothermal tempering curves for* 0·14%C–12%Cr *steel*

a single curve is obtained representative of a wide range of different tempering conditions. The use of this method can be justified for the short tempering times used (1–100 h) although for steels which have a strong secondary hardening reaction it is necessary to check the main features with isothermal tempering curves.

The tempering curve for the base composition is shown in Fig.4*a*. The tempering temperatures of the individual points have been identified and it will be seen that all of these fall on to a smooth curve. The times of tempering at each temperature were 1, 2, 4, 7, 10, 25, 50, and 100 h. (In later tempering curves the abscissa also identifies the parameter which represents the 1 h treatment at different temperatures, for convenience in interpretation.) Since the steel is air hardening the initial hardness represents the martensitic hardness of the 0·1%C–12%Cr steel, which is higher than is usual for a 0·1%C low-alloy steel. There are two main reasons for this:

(i) increasing Cr causes some solid solution hardening

(ii) 12%Cr causes a marked depression of the M_s from 450°C for low-alloy steels to 300°C for 12%Cr steels. This decreases the amount of autotempering and causes the carbon to be made fully effective in hardening the martensite.

Both these effects contribute to the martensitic hardness, and when other alloying elements are added to the 12%Cr base they too act in a similar way, although there is probably less effect in suppressing the autotempering when the M_s falls below 300°C. In this case, more of the effect is due to solid solution hardening, which is retained even at the highest tempering temperature. When the tempering curves of the alloyed 12%Cr steels, which are described later, are examined, it is possible to see the separate effect of the different alloying elements on the martensite hardness.

The first effect of tempering is a decrease in the hardness level and a minimum is reached at about 300–350°C. Slight secondary hardening then occurs which reaches a maximum at 450°C. It should be emphasized that these effects are slight however, as the following figures show:

Condition	Hardness, DPN
Solution treated	455
T 350°C /2h	450
T 450°C /75h	460

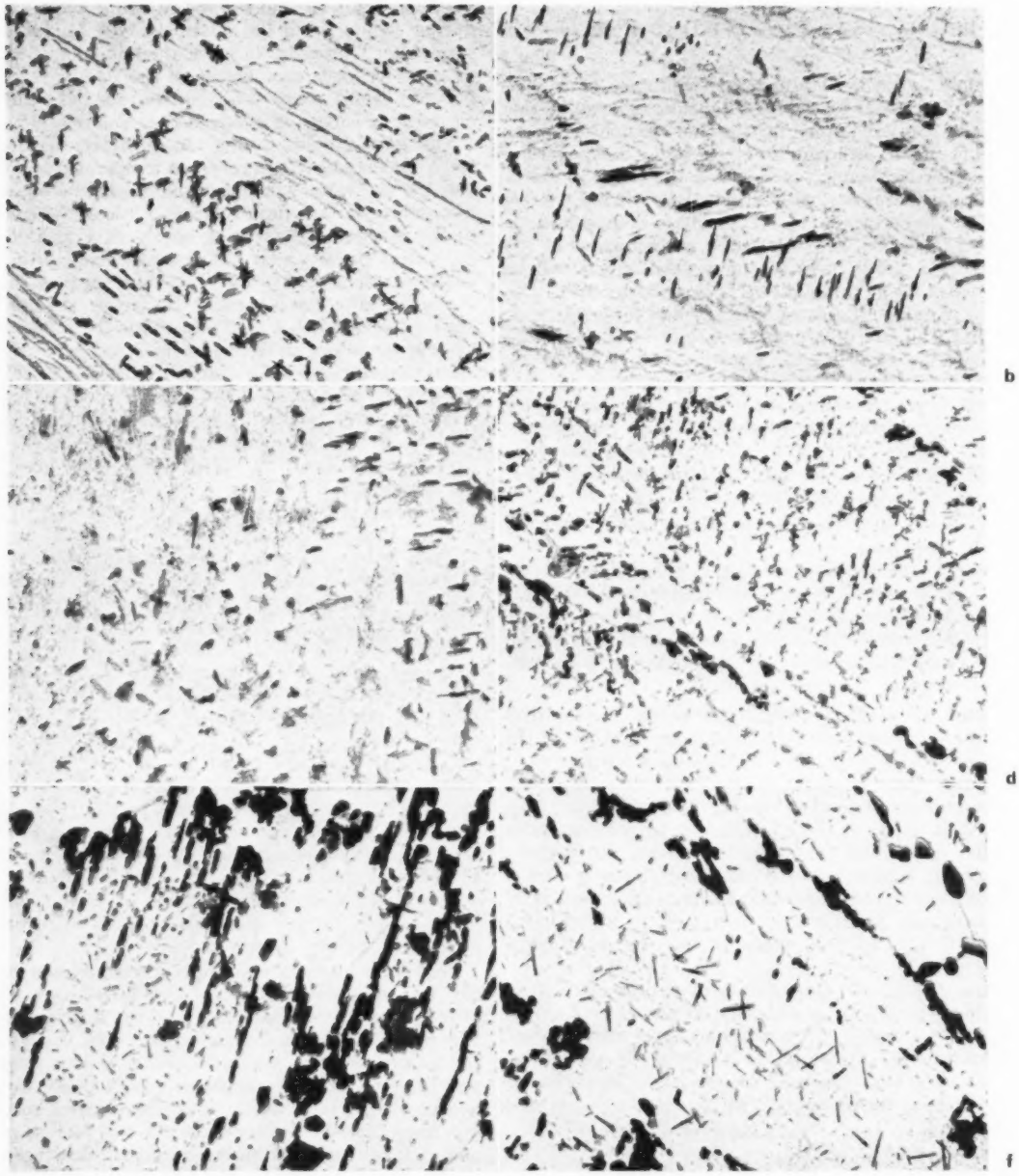
Overageing occurs quite rapidly above 500°C and the hardness drops to about 250 DPN after 1 h tempering in the range 600–700°C. These effects can also be seen from the isothermal curves shown in Fig.4*b*.

It may help the subsequent description of the effect of alloying elements if the structural changes during tempering the base 12%Cr steel are related to the tempering curve. It will then be possible to assess the effect of alloying elements by considering the differences between alloyed steels, and the base composition. The solution-treated structure can be considered as a typical untempered martensitic structure. The martensitic plates or needles can be seen and there may be small particles of Fe_3C precipitated within them. These result from slight autotempering which occurs during cooling from the relatively high M_s temperature. After tempering at 300°C the only change in the microstructure is an increase in the number of Fe_3C particles which presumably causes the slight drop in hardness as carbon is removed from solution (Fig.5*a*). A change is noticed during the secondary hardening stage of the tempering process at temperatures of 400° and 450°C. This is where the combined time/temperature parameter has limitations and it is useful to refer to the isothermal tempering curves in Fig.4*b*. These show that the secondary hardening occurs with long heating periods at 300–400°C and the hardness is still increasing after 100 h. At 450°C however, the hardening process reaches a maximum after 10 h and then overageing takes place. There is evidence of a fine matrix precipitate at 450° and 500°C (Figs.5*b* and *c*) but this can be seen much more clearly in the overaged condition after tempering for longer times at 500°C (Fig.5*d*). It will be noted that the original Fe_3C precipitate is re-dissolving during the formation of the fine matrix precipitates.

This fine precipitate appears to be mainly Cr_7C_3 which is formed by separate nucleation, but there also appears to be a little M_2X phase present, which is presumably based on Cr_2C. Apart from the separately nucleated precipitate there is a transformation from Fe_3C to Cr_7C_3 *in situ* in the Fe_3C particles, and this causes small rods of Cr_7C_3 to take the place of the Fe_3C as shown in Fig.5*d*. The nature of the M_2X phase has already been discussed elsewhere[5–7] and it is sufficient to say that in plain 12%Cr or 17%Cr steels it undoubtedly is based on Cr_2C. Increasing the N content or adding alloying elements such as Mo or V, or for that matter many other alloying elements, stabilizes the M_2X in preference to the Cr_7C_3 and in 12%Cr–Ni–Mo–steels M_2X is the major secondary hardening precipitate. Because of its structure, M_2X can dissolve large amounts of many alloy elements, and thereby can cause extra secondary hardening and also be stabilized over a wide temperature range, with a resultant increase in the tempering resistance.

Secondary hardening can be increased by increasing the lattice parameter of the precipitating carbide relative to the lattice parameter of the ferrite matrix. For non-carbide-forming elements which dissolve in the matrix (silicon, aluminium, nickel, cobalt, copper, and to some extent, manganese), only silicon would be expected to have much effect in increasing secondary hardening since this decreases the lattice parameter. The carbide-forming elements increase the lattice parameter of the carbide in the increasing order vanadium, molybdenum, tungsten, titanium, niobium, and therefore will be expected to increase the amount of secondary hardening in the same order.

At the same stage in tempering, namely after tempering at 500°C, the formation of relatively large carbide particles is observed at martensite plate and austenite grain boundaries (Fig.5*d*). It was possible to identify these both by electron diffraction of replicas and by X-ray diffraction of extracted residues as $M_{23}C_6$. As tempering proceeds, this boundary precipitate grows at the expense of the fine Cr_7C_3 precipitate, which coarsens (Fig.5*e*), and it is clear that this is the structural change which results in the pronounced softening shown by the tempering curve. Further growth of the $M_{23}C_6$ occurs on extended tempering with more solution of the Cr_7C_3 but it is seldom that any grain growth of the ferrite occurs at temperatures up to 700°C (Fig.5*f*). Consequently, above 600°C the steel has lost practically all of the hardness which was associated with the martensitic structure and subsequent secondary hardening. It was the major objective of this work to improve the poor tempering resistance of the base 12%Cr steel by alloy additions.



Tempered at *a* 300°C for 1 h ×15000 *b* 450°C for 240 h ×15000 *c* 500°C for 8 h ×15000 *d* 500°C for 240 h ×15000 *e* at 600°C for 8 h ×15000 *f* at 700°C for 1 h ×15000

5 *Electron micrographs of structures produced during tempering a* 0·1%C–12%Cr *steel*

EFFECT OF AUSTENITE-FORMING ELEMENTS

Effect of carbon

Carbon additions of between 0·03% and 0·21% were made to the base composition and the tempering curves obtained are shown in Fig.6*a*. The most important effect is an increase in the martensitic hardness level and the solution-treated hardness values are given below:

% Carbon	Hardness, DPN
0·029	261
0·068	364
0·133	462
0·206	480

Apart from the effect on the martensitic hardness level carbon also increases the amount of secondary hardening which occurs at 450°C. If the amount of secondary hardening is measured as the difference between the minimum and maximum hardnesses reached during tempering between 300° and 500°C then the effect of carbon can be summarized as follows:

% Carbon	Hardness change, DPN
0·029	+9
0·068	+11
0·133	+41
0·206	+41

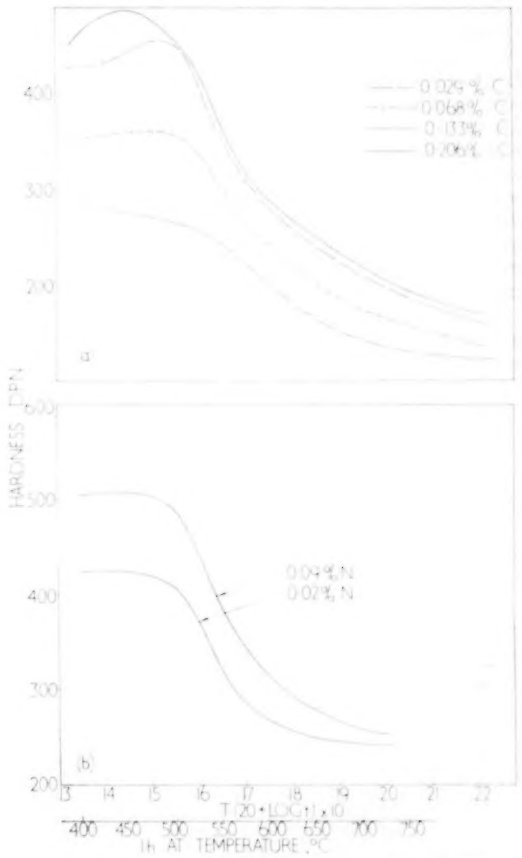
The third important effect is the hardness difference produced after tempering at 650°C, which would be an appropriate tempering temperature for commercial steels. It will be seen that the effect of carbon has decreased very much, particularly above 0·15%C.

% Carbon	Hardness after tempering 1 h at 650°C, DPN
0·029	172
0·068	206
0·133	235
0·206	250

The effect of carbon on the microstructure after tempering 1 h at 650°C is shown in Figs.7*a* and *b*. It will be seen that there is an increase in the amount of the matrix Cr_7C_3 and M_2X precipitates and this is responsible for the hardness increase. The effect is not very great, however, and apart from the fact that high hardnesses can be obtained at low tempering temperatures, there is little justification for using high carbon contents. Other work has shown that the ductility and impact properties are reduced very much at high carbon levels, so it would seem that high-carbon steels will only find application where hardness alone is the most important property.

Effect of nitrogen

Steels with the base composition were made with nitrogen levels of 0·02, and 0·09% and the tempering curves are shown in Fig.6*b*. Nitrogen has a similar effect to carbon and both the martensitic hardness and the amount of secondary hardening are increased. As with carbon, the effect of nitrogen is much reduced at higher tempering parameters. Microstructures representing the structure after tempering 1 h at 650°C are shown in Figs.7*b* and *c* (these are 0·14%C steels with 0·02 and 0·09%N) and it will be seen that increasing nitrogen intensifies the fine matrix precipitate which changes from Cr_7C_3 to mainly M_2X as already described. This M_2X phase, which was identified by



a effect of carbon in a 0·02%N steel *b* effect of nitrogen in a 0·1%C steel

6 *Effect of carbon and nitrogen on the tempering curve of a* 12%Cr *steel*

electron diffraction of extraction replicas, could either be a carbide or nitride and in this case may well be based on Cr_2N.

The effect of nickel

Nickel additions up to 5% were made to the base composition and the tempering curves obtained are shown in Fig.8*a*. It is convenient to consider the effect of each alloying element on the martensitic hardness, the amount of secondary hardening and the hardness level after tempering at 650°C for 1 h. These effects are summarized below:

% Ni	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h 650°C, DPN
0	415	+10	230
1·04	465	+15	280
2·15	440	+10	250
3·03	480	+10	290
4·24	470	+20	285
5·22	508	+40	305

The secondary hardness change is the difference between the minimum hardness tempered at 350°C and the maximum hardness at 400–450°C. Nickel increases the martensitic hardness and this must be due to both

solid solution hardening and its effect in depressing the M_s temperature, which produces less autotempering. The M_s actually decreases to about 175°C with 5%Ni. Nickel also slightly increases the amount of secondary hardening and it will be noticed that this effect occurs at a slightly lower tempering parameter. This produces some acceleration in the rate of softening, but the tempering curves then cross over at higher tempering parameters which shows quite clearly that increasing nickel is retarding the final stages of softening. The increase in hardness of the high-nickel steels when tempered at 650°C shows that the nickel addition has depressed the A_{c1} temperature below 650°C.

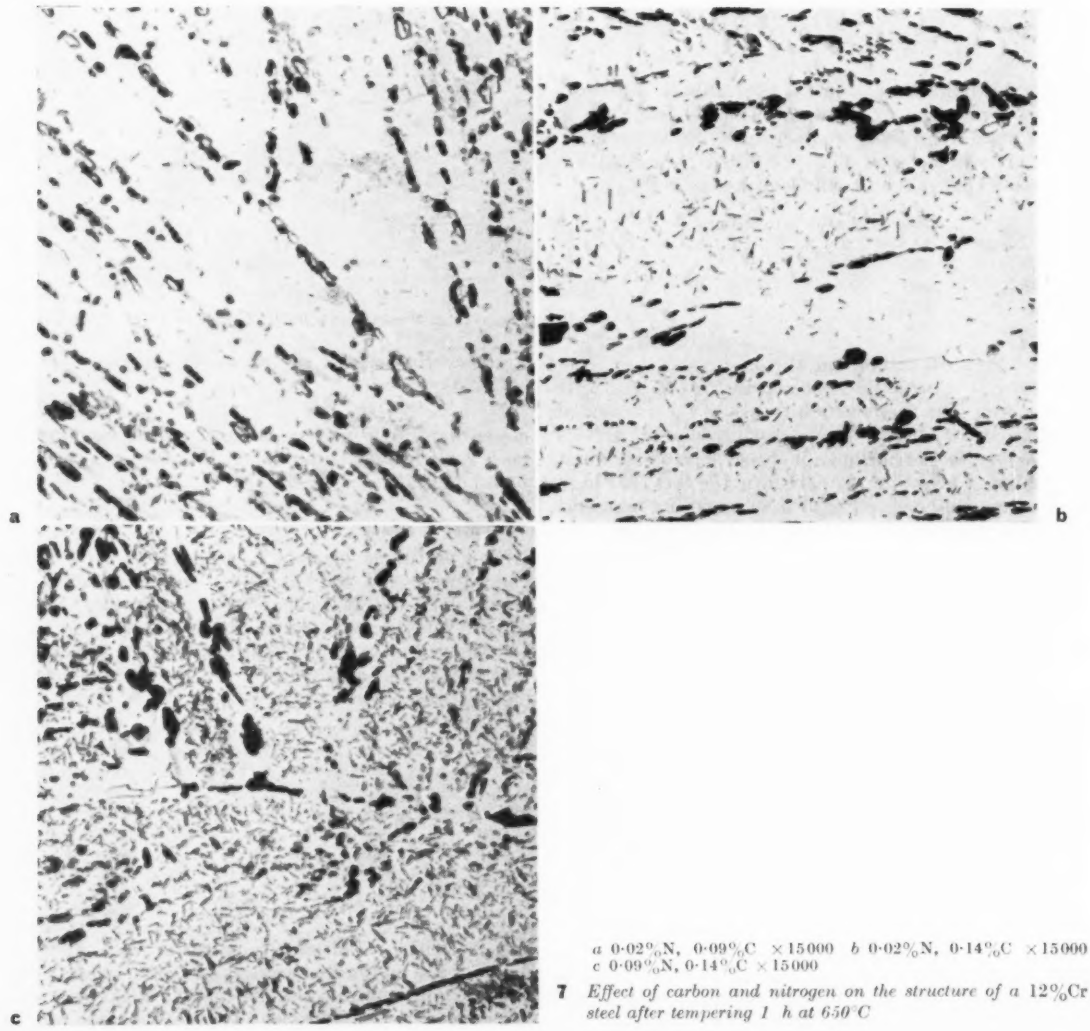
A detailed electron microscope study of these steels showed the same tempering process as the base steel previously described. The carbide change is from

$$Fe_3C \rightarrow \left\{ \begin{matrix} M_2X \\ + \\ Cr_7C_3 \end{matrix} \right\} \rightarrow M_{23}C_6.$$

A comparison of the 3%Ni and nickel-free steels is shown in Fig.9 after tempering at 500°C for 8 h which produces the peak of secondary hardening. Figure 9*b* clearly shows the increased amount of precipitatewhich is producing the slightly increased secondary hardening. It is perhaps surprising that nickel should increase the amount of carbide since it is not a carbide-forming element. The most likely explanation is that nickel decreases the solubility of carbon in the matrix and hence leads to a greater supersaturation effect which promotes the formation of the Cr_7C_3 and M_2X carbide precipitates. This same effect of nickel producing supersaturation is also responsible for the acceleration of the secondary hardening process. At higher tempering temperatures the higher Ni steels again show greater tempering resistance. In part this is due to re-austenitization and in part to solid solution hardening. Also, a contributory factor may be the slower diffusion in the Ni steels which slightly retards both carbide solution and grain growth.

Effect of manganese

Manganese additions up to 4% were made to the base composition and the tempering curves obtained are



a 0·02%N, 0·09%C ×15000 *b* 0·02%N, 0·14%C ×15000
c 0·09%N, 0·14%C ×15000

7 *Effect of carbon and nitrogen on the structure of a* 12%Cr *steel after tempering 1 h at 650°C*

shown in Fig.8*b*. The most important effects are summarized below:

% Mn	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h 650°C, DPN
0·58	415	+10	230
1·03	404	+10	235
2·27	415	+15	232
4·28	464	+25	270

It will be seen that manganese has a similar effect to nickel and the same explanations can be given for these effects. There is less justification for using manganese as an alloying element however, since it is not as strong an austenite-forming element as nickel.

Effect of copper

Copper additions up to 4% were made to the base composition and the tempering curves obtained are shown in Fig.8*c*. The main effects are summarized below:

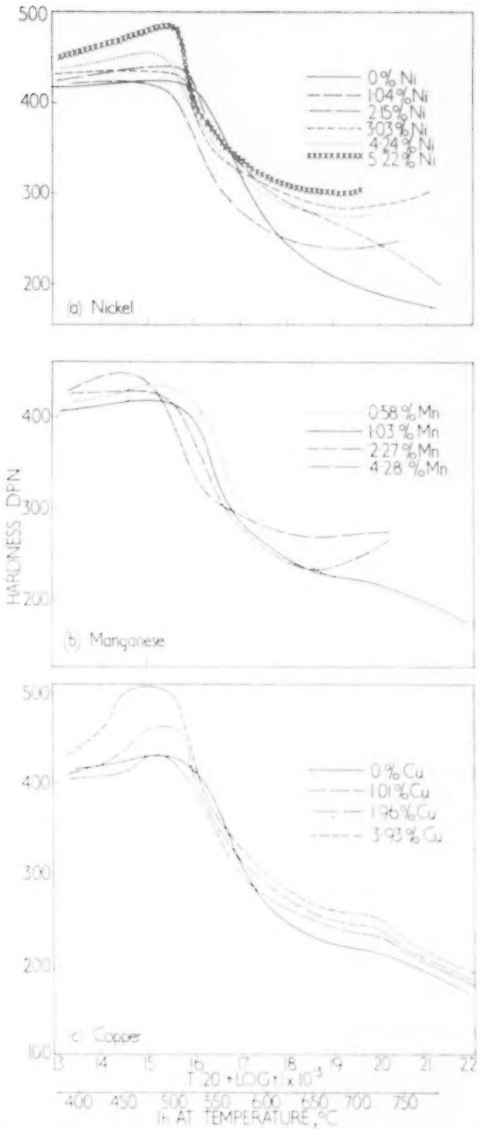
%Cu	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h 650°C, DPN
0	415	+10	230
1·01	413	+30	253
1·96	430	+53	262
3·93	425	+80	272

Copper has little or no effect on the martensitic hardness but a very marked effect on the secondary hardening reaction. This is accompanied by a slight acceleration of the process so that overageing occurs more rapidly. In the later stages of tempering however, the higher copper steels retain a higher hardness level. It was not possible to identify phases other than the carbides which were formed in the base composition, and consequently it was concluded that the secondary hardening was mainly due to the formation of M_2X. Typical microstructures showing the precipitation after tempering for 1 h at 650°C are given in Fig.10, which show that the increased secondary hardening is due to an increase in the amount of the fine matrix precipitate of M_2X although the formation of a copper-rich precipitate cannot be completely ruled out. The increased hardness of the higher copper steels at high tempering parameters is presumably due to solid solution hardening and the greater number of precipitated particles as shown in Fig.10.

Effect of cobalt

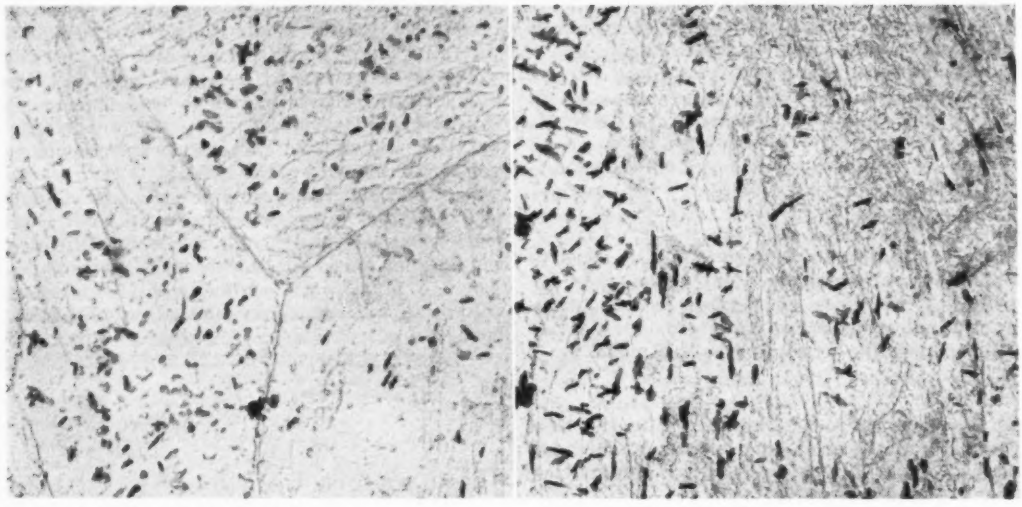
Cobalt additions up to 10% were made to the base composition and the tempering curves obtained are shown in Fig.11*a*. Such large additions of cobalt were possible because of the slight effect it has in depressing both the A_{c1} and the M_s temperature. The main effects of cobalt can be summarized as follows:

%Co	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h 650°C, DPN
0	415	+10	230
1·01	429	+15	255
1·98	466	+20	260
4·12	452	+10	263
7·15	496	+10	275
10·0	505	+10	285



8 *Effect of alloying elements on the tempering resistance of* 0·1%C–12%Cr *steels*

Cobalt increases the martensitic hardness owing mainly to solid solution hardening and slightly to its effect in depressing the M_s temperature. This hardness increase is maintained up to 500°C but there does not appear to be any pronounced secondary hardening. The hardness increase is maintained at high tempering parameters presumably owing to the solid solution hardening. A final point of interest is that cobalt does not have much effect in depressing the A_{c1} temperature since even with 10%Co present the A_{c1} temperature is still above 700°C. Cobalt can be considered as a useful alloying element apart from its high cost. It is an austenite-forming element which produces a hardness increase at all stages of tempering and yet it has only a slight effect in depressing the A_{c1} and M_s temperatures.



a 0% nickel × 15000 *b* 3% nickel × 15000

9 *Effect of nickel on the structure of a 12%Cr steel after secondary hardening by tempering at 500°C for 8 h*

THE EFFECT OF FERRITE-FORMING ELEMENTS

It has been mentioned previously that the 12%Cr base composition is near to the borderline of forming ferrite during solution treatment. Consequently, it is only possible to study the effect of appreciable additions of ferrite-forming elements if the base composition is moved further into the stable austenite region at solution temperature by the addition of an austenite-forming element. The possibilities are C, N, Ni, Mn, Co, and Cu and the effect of each of these has been described in the previous section. Carbon and nitrogen both have an adverse effect on ductility, impact resistance, and weldability, and their use must be restricted. Manganese and cobalt are relatively weak austenite-formers and in addition, cobalt is expensive. Nickel is the most suitable element to use but even so its use must be limited because of the marked effect it has in depressing the A_{e1} and M_s temperatures. A reasonable maximum is 3%Ni but in order to allow for the effect of other alloying elements especially on the M_s temperature it was decided to limit nickel to 2%. This has the advantage that it also allows some slight increase in nickel content of any selected alloy if this is found to be necessary to retain a ferrite-free structure up to 1300°C compared with the more usual solution temperature of 1050°C. The modified base composition which was chosen for the work on the effect of ferrite-forming elements was therefore 0·1%C, 12%Cr, 0·5%Mn, 0·25%Si, 2·0%Ni, 0·02%N. The only exception was in the molybdenum series where a high N (0·09%) was used in order to illustrate a specific effect.

Effect of molybdenum

Molybdenum additions up to 4% were made to the 2%Ni base composition which contained 0·09%N. It will be shown that one of the most important effects of molybdenum is to stabilize the M_2X precipitate. This can be seen most easily when the amount of the M_2X precipitate is increased by using a high nitrogen content. The effects shown for increasing molybdenum are also obtained in low-nitrogen steels but at a lower hardness level as indicated by the low-nitrogen base steel which was included in this series. The tempering curves obtained are shown in Fig.11*b* and the main effects are summarized below:

%Mo	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h 650°C, DPN
0 (low N)	430	+10	240
0 (high N)	531	+10	280
1·0	530	+12	243
2·04	537	+20	300
3·10	523	+30	335
4·07	512	+15	360

All of the steels were ferrite-free except for the 4%Mo steel which contained 7% ferrite. Molybdenum has little effect in raising the martensitic hardness and this may be due to the retention of some austenite. The secondary hardening reaction is intensified at least up to 3%Mo and there is a marked increase in hardness at the higher tempering parameters. Electron microscope examination showed that the secondary hardening effect was due to a fine matrix precipitate and the microstructures after tempering 1 h at 650°C are shown in Fig.12. These show how the M_2X precipitate is refined and stabilized by molybdenum so that it can be present as a fine dispersion at high tempering parameters. This effect, together with some solid solution hardening resulted in the increased hardness.

Effect of tungsten

Tungsten additions up to 3% were made to the 2%Ni base composition and the tempering curves obtained are shown in Fig.13*a*. It should be noted that all of these steels were ferrite-free. The main effects of tungsten are summarized below:

%W	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h 650°C, DPN
0	430	+10	240
0·54	475	+12	280
1·01	490	+15	285
1·90	490	+10	285
2·99	505	+20	312

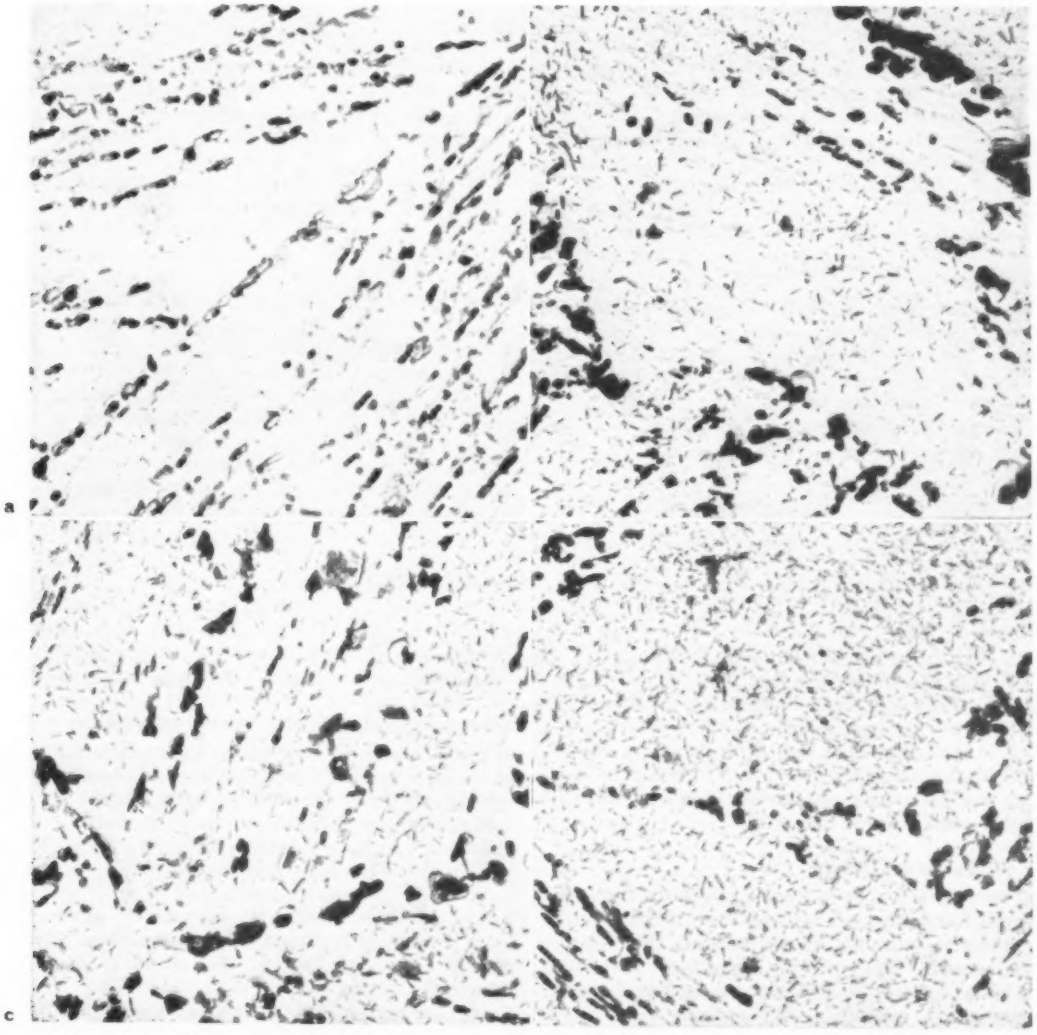
Tungsten had a similar effect to molybdenum on the microstructures and tempering curves. The secondary hardening effect was increased owing to the intensification of the M_2X precipitate, and a higher hardness was maintained at high tempering parameters owing to the stabilization of the M_2X precipitate and solid solution hardening. Tungsten was less effective than molybdenum since almost twice as much was required to produce an equivalent effect.

Effect of vanadium

Vanadium additions up to 1·78% were made to the 2%Ni base composition and the tempering curves obtained are shown in Fig.13*b*. It should be noted that the 1·78%V steel contained 13% ferrite which has affected the tempering curve of this steel. It can be concluded that little more than 1%V can be added to this base composition without producing ferrite. The main effects of vanadium are summarized below:

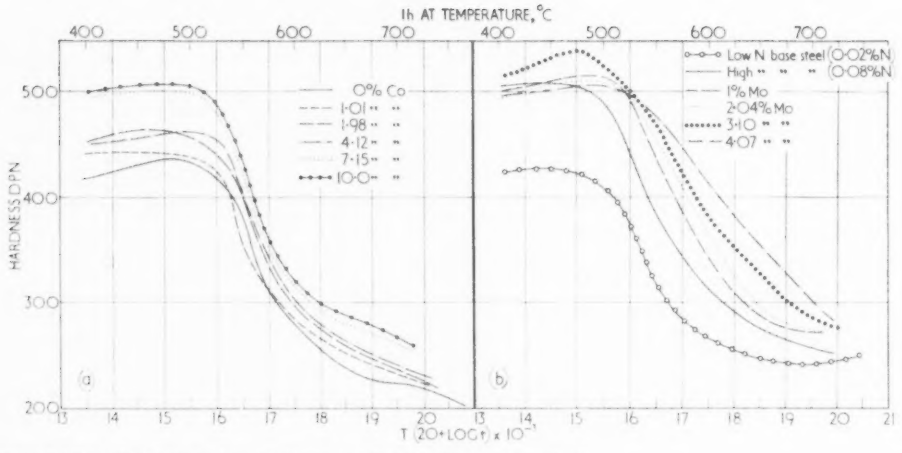
%V	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h at 650°C, DPN
0	430	+10	240
0·9	490	+10	325
1·78	417 (13% ferrite)	+5	280

Vanadium raises the hardness level at all stages of



a 0%Cu ×15000 *b* 1·01%Cu ×15000 *c* 1·96%Cu ×15000 *d* 3·93%Cu ×15000

10 *Effect of copper on the structure of a* 0·1%C–12%Cr *steel after tempering at 650°C for 1 h*



a cobalt in a 0·1%C–12%Cr steel *b* molybdenum in a 0·1%C–12%Cr–2%Ni steel

11 *Effect of cobalt and molybdenum on the tempering resistance of* 12%Cr *steels*

tempering and this is particularly noticeable between 550° and 650°C due to the stabilization of the M_2X precipitate which produces the secondary hardening. The introduction of ferrite in the 1·78%V steel however, caused the hardness to be low, and also the intensity of the secondary hardening to be less. No V_4C_3 was observed in the tempered martensite, but occasionally V_4C_3 plates were seen in the delta ferrite.

Effect of silicon

Silicon additions up to 2·8% were made to the 2%Ni base composition. The tempering curves obtained are shown in Fig.13*c*. It should be noted that these steels were ferrite-free apart from the one containing 2·8%Si which had 6% ferrite in the microstructure after solution treating at 1050°C. The main effects are summarized below:

%Si	Soln. treated hardness, DPN	Secondary hardening change, DPN	Hardness after tempering 1 h 650°C, DPN
0·24	430	+10	240
0·79	470	+5	280
1·94	481	+15	298
2·80	475	+10	315

Silicon slightly raises the martensitic hardness and this increase is maintained after tempering. It will be seen that there is a small increase in the secondary hardening due to Si contracting the ferrite lattice but this effect is smaller than can be produced by dissolving carbide-forming elements in M_2X. The higher hardness at high tempering parameters is presumably due mainly to solid solution hardening.

Effect of titanium and niobium

These two elements have similar effects in that they are strong ferrite-formers and also react with carbon to form stable carbides and therefore lower the effective carbon content. Consequently, they have a marked effect both on the initial martensitic hardness and on the tempering resistance. While it is possible to dissolve these carbides using very high solution temperatures, this would seriously limit the commercial use of such steels, therefore, it was decided to use the standard solution temperature of 1050°C for this work.

An addition of ½%Ti was made to 12%Cr steels containing 1 and 3%Ni and the tempering curves obtained are shown in Fig.14. With 1%Ni, the addition of ½%Ti introduces 29% ferrite and the effect of this can be seen in the low initial martensitic hardness shown in Fig.14*a*. The fact that the two tempering curves approach one another at high tempering parameters indicates that there has been some retardation of tempering. The ferrite is eliminated with 3%Ni, but Fig.14*b* shows that the initial hardness level is still low owing to carbides being out of solution. It will be seen that definite secondary hardening is now giving improved tempering resistance.

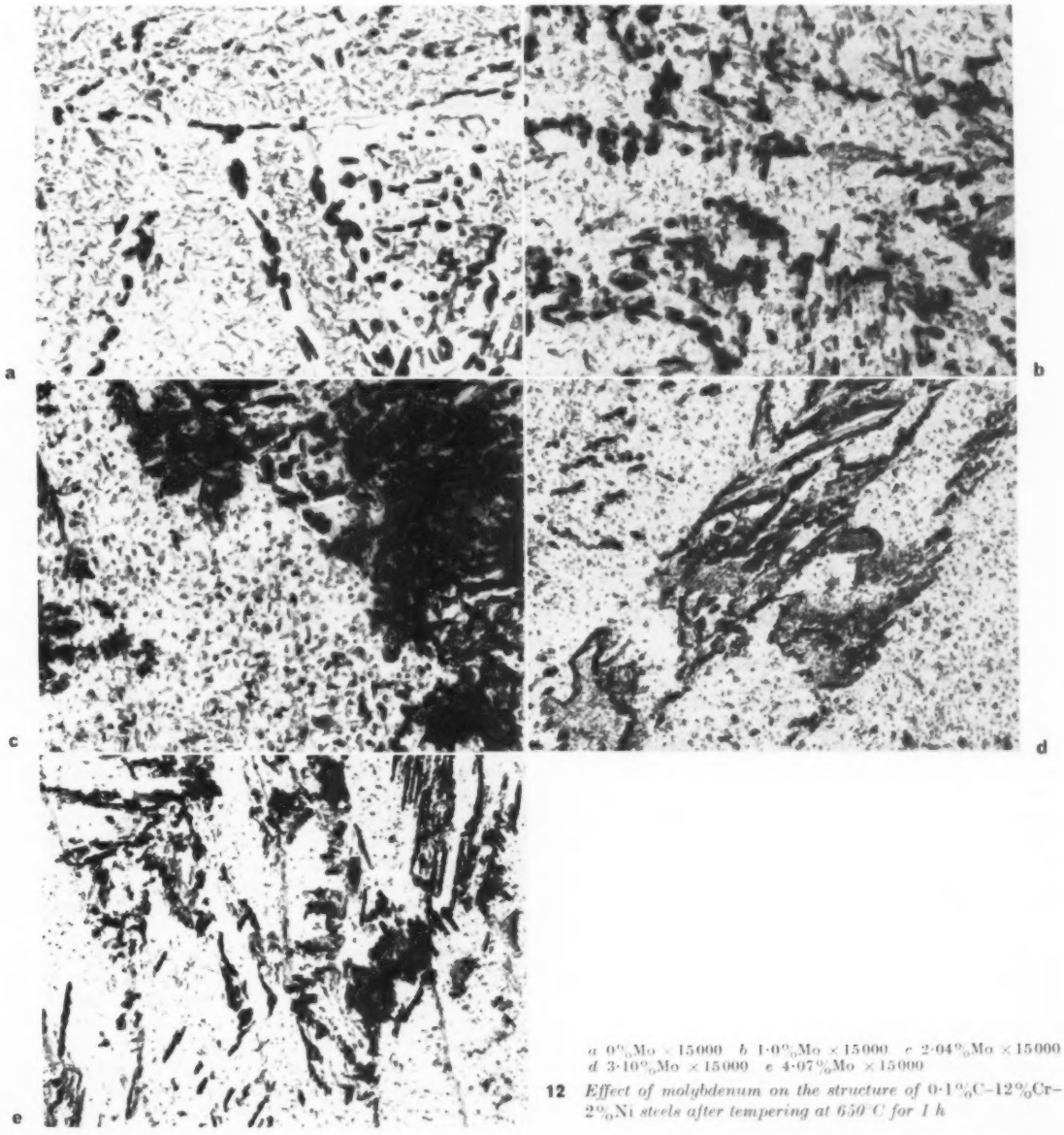
The similar effects of Nb in lowering the martensitic hardness level owing to carbide formation and also promoting ferrite formation are shown in Fig.14*c* for 12%Cr–2%Ni–Mo steels. The effect of solution treatment temperature is also shown. Despite starting from a lower initial hardness value the Nb steels have appreciably better tempering resistance between 550° and 700°C. Increasing the solution temperature to 1200°C further increased the tempering resistance and caused more intense secondary hardening. Both Ti and Nb dissolve in the precipitating M_2X carbide to give an expanded lattice parameter. The M_2X carbide is also stabilized to higher tempering temperatures.

Effect of alloying elements on mechanical properties

While the main interest in developing alloyed 12%Cr steels is to produce increased tempering resistance it is necessary to have adequate impact strength associated with the tensile strength, and for the tensile properties themselves to be satisfactory with a 0·2%PS/TS ratio of 0·75 to 0·80 and adequate ductility. The results of a considerable amount of work can be summarized by saying that these properties depend very much on the stage of tempering or the microstructure which is obtained and also upon the effect of individual alloying elements.

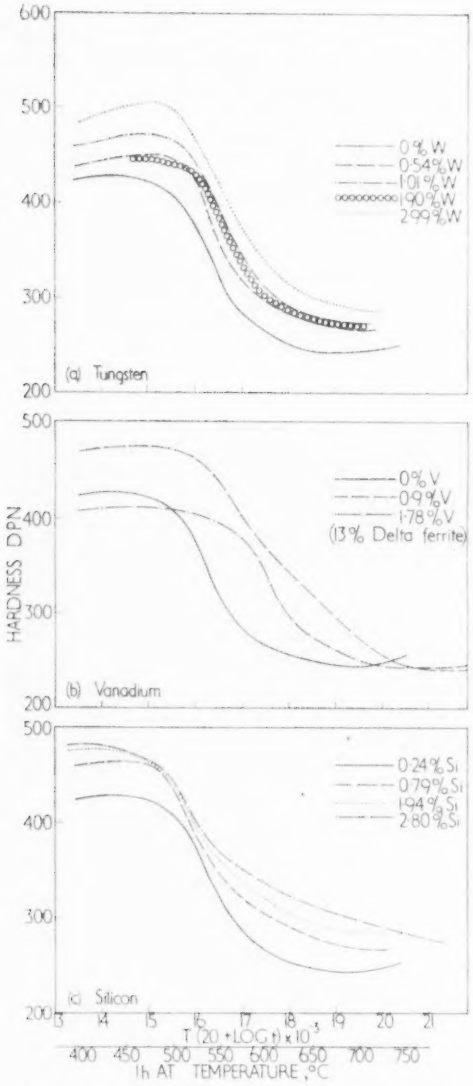
It is appropriate first to consider the impact properties. There is a characteristic curve which shows the change in room temperature impact value with increased tempering temperature for various 12%Cr

steels and this is shown in Fig.15*a*. These room temperature energy values are of course related to the position of the impact transition curve and the full transition curves for one of the steels included in Fig.15*a* are shown in Fig.16. It is simpler, however, to examine the effect of an alloying element in terms of the room temperature impact value. The impact value of the base 12%Cr-2%Ni steel in the untempered or very lightly tempered condition is low but is raised by tempering at 300°C. This is due to the relief of internal stresses in the martensitic structure which is accentuated by the precipitation of some Fe_3C (as described previously and shown in Fig.5). Reasonably good impact values are obtained with tempering temperatures between 200° and 400°C, but in the upper part of this tempering range, the impact value is dropping and continues to drop until it reaches a minimum with tempering at about 550°C. Above this temperature the energy value begins to increase (it will be remembered that the tensile strength is decreasing during this stage of tempering) but it will be noted that the 12%Cr–2%Ni steel does not improve nearly so much as the steels containing molybdenum. The explanation for this trough in the impact energy curve can be found in the sequence of microstructural changes. Between 400° and 500°C there is a secondary hardening process occurring which is producing coherency strains and resistance to deformation. Overageing is also occurring at grain boundaries, however, so that a much coarser precipitate is forming together with local alloy and carbon denudation. While it is not clear how much this denudation contributes to the effect it would appear

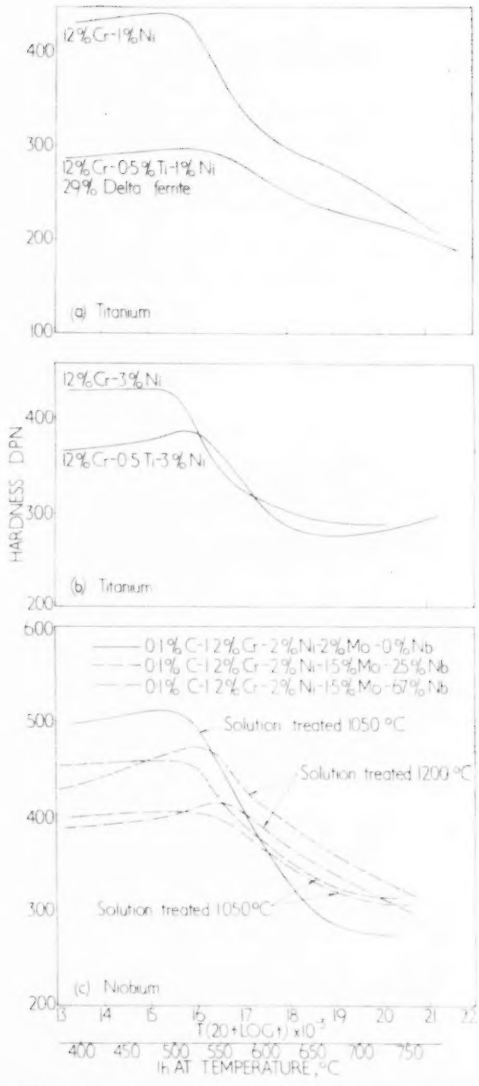


a 0%Mo ×15000 *b* 1·0%Mo ×15000 *c* 2·04%Mo ×15000 *d* 3·10%Mo ×15000 *e* 4·07%Mo ×15000

12 *Effect of molybdenum on the structure of* 0·1%C–12%Cr–2%Ni *steels after tempering at 650°C for 1 h*



13 *Effect of alloying elements on the tempering resistance of* 0·1%C–12%Cr–2%Ni *steels*

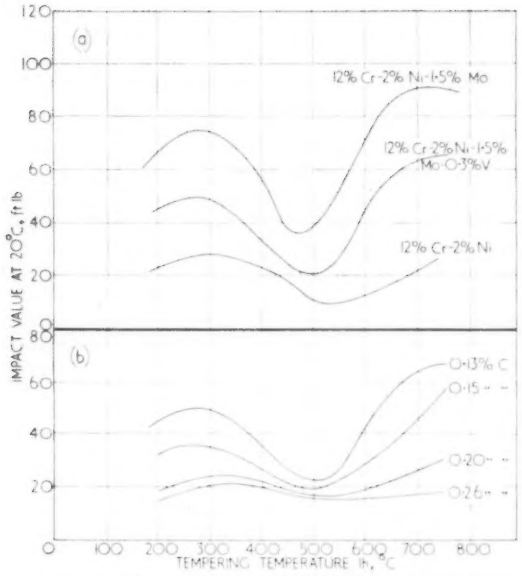


14 *Effect of titanium and niobium on the tempering resistance of* 0·1%C–12%Cr–Ni *steels*

that the local overageing at the grain boundaries leads to crack initiation and the subsequent propagation of brittle failure through a structure which is not very crack resistant. It might be expected that an element like Mo which strengthens the ferrite would tend to inhibit crack initiation. This is confirmed by the improved impact properties of the 1½%Mo steel shown in Fig.15*a*. While there is still a trough in the curve, the energy values are higher throughout the tempering range and quite reasonable even at the bottom of the trough. In fact, the only two elements which have been found to have a marked beneficial effect are molybdenum and tungsten and it would appear that 1·25–1·5%Mo and 2·5–3·0%W are necessary to obtain this improvement. If the Mo is raised too far, however, the increased secondary hardening begins to affect the impact properties and this applies, for instance, to 3 and 4%Mo steels. Similarly vanadium also intensifies and stabilizes the secondary hardening M_2X precipitate and this adversely affects the impact properties as shown in Fig.15*a*.

While it is believed that it is the secondary hardening process which is controlling the impact properties, another factor which cannot be ruled out is some degree of temper embrittlement. Even if this were present, however, molybdenum would still have a beneficial effect.

Carbon is known to have an adverse effect on the impact properties of steels, decreasing the energy value which can be obtained even with a fully ductile fracture. The effect of carbon on the impact value of an alloyed 12%Cr steel is shown in Fig.15*b*. While the effect is slight up to 0·15%C it becomes more pronounced above this and rather low impact values are obtained with 0·20 and 0·25%C. A very similar effect



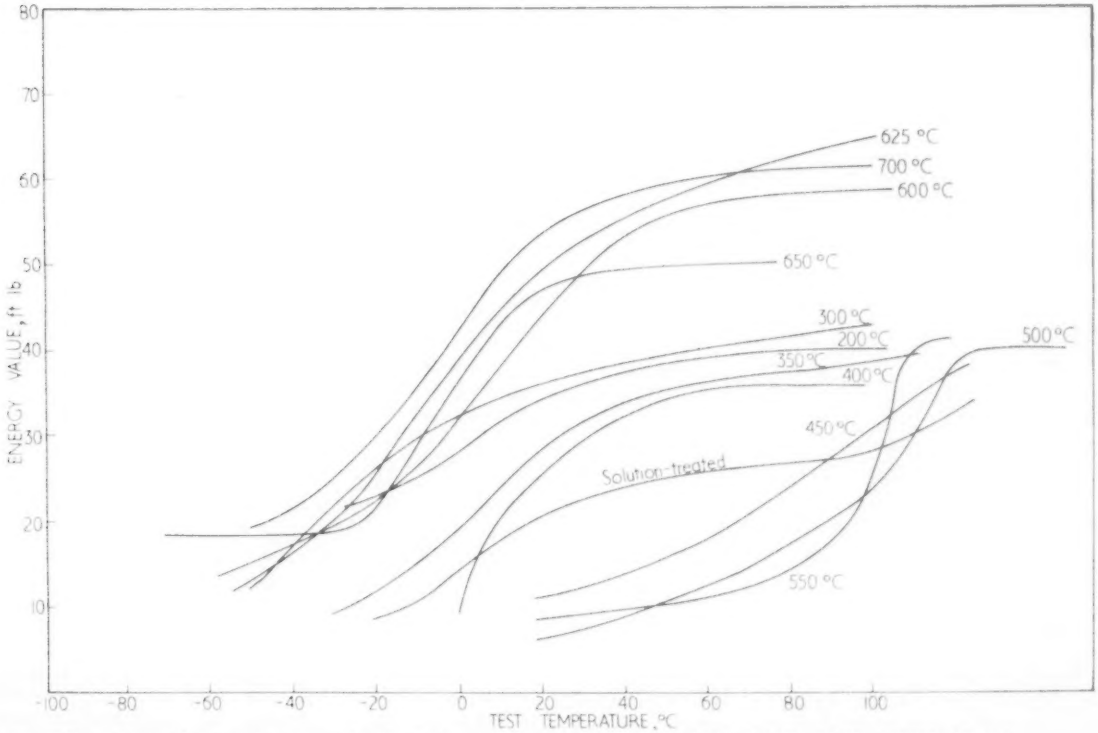
a effect of Mo and V *b* effect of C (12% Cr, 2% Ni, 1·5% Mo, 0·3% V steels)

15 *Effect of tempering on impact properties of* 12%Cr *steels*

was observed with nitrogen and it has been shown that the impact values which can be obtained with 0·01–0·04% nitrogen are very much better than those which can be obtained with 0·06–0·08% nitrogen.

With regard to the tensile properties it is most important to have satisfactory proof stress values associated with high tensile strengths. It is known that, in general, tempered martensitic structures have satisfactory properties with a 0·2%PS/TS ratio of about 0·75–0·80, and this applies equally well to the base 12%Cr steel. There is a complication in this case because of the secondary hardening reaction which occurs on tempering at 450–500°C. This introduces coherency strains which tend to lower the proof stress values in spite of the fact that the hardness and tensile strength are increased. The change in tensile properties with tempering for a typical 0·13%C–12%Cr steel are shown in Fig.17. It will be seen that the change in tensile strength is similar to that shown in hardness tempering curves in that the secondary hardening process produces an increase in strength at a tempering temperature of 550°C. The 0·2% proof stress does not follow the same curve and begins to decrease relative to the tensile strength (i.e. the 0·2%PS/TS ratio decreases) when this secondary hardening occurs. The effect is most noticeable at the peak of secondary hardening. When overageing occurs the coherency stresses are relieved and the PS/TS ratio increases again reaching maximum values with tempering between 600–650°C.

It has been mentioned that one of the major objectives in this work was to improve the tempering resistance of 12%Cr steels. Because of the low proof stress values and poor impact properties associated with the peak of secondary hardening, the main interest is in the properties which can be obtained after tempering at 650°C, when both the proof stress values and the impact strength are high. It can be



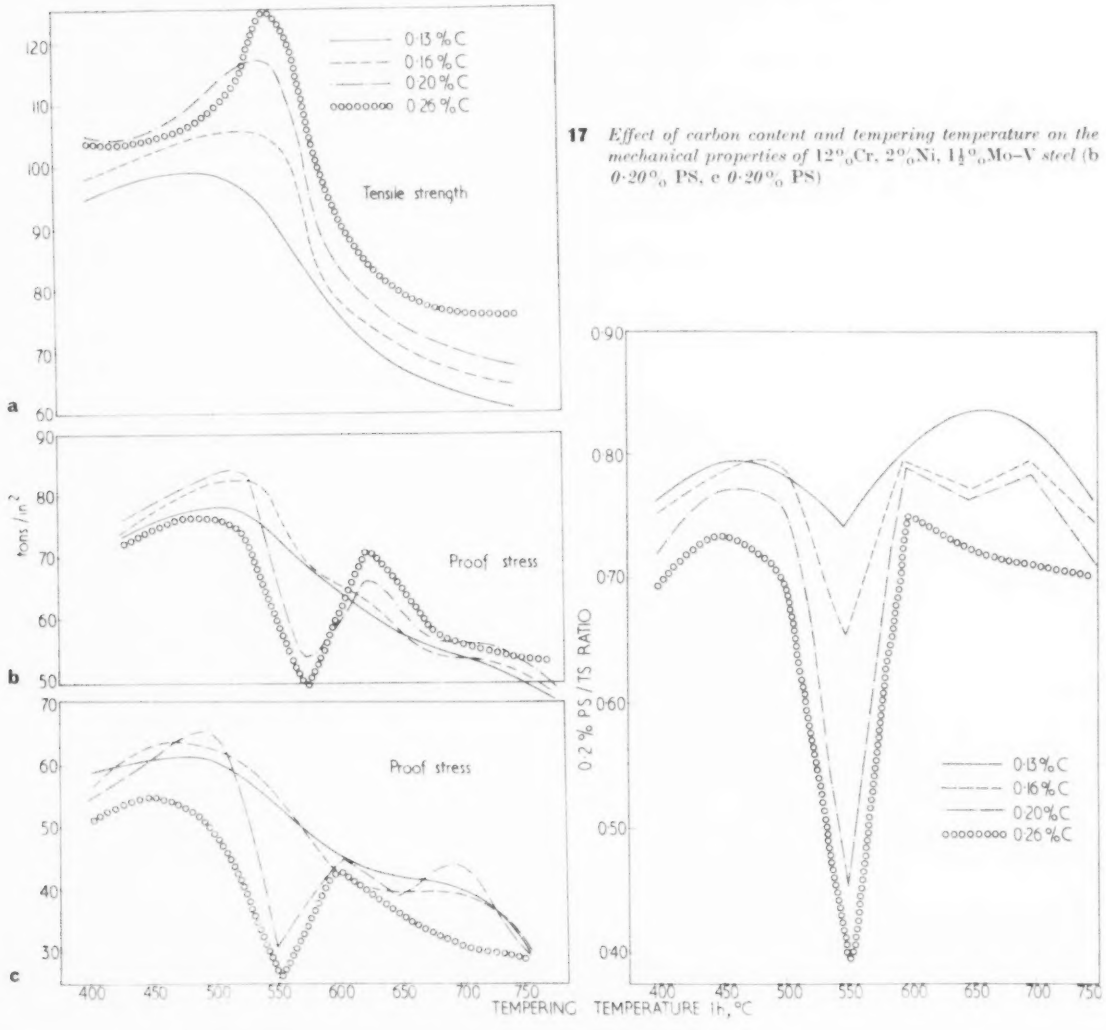
16 *Effect of tempering temperature on the Charpy impact transition curves of* 0·1%C, 12%Cr, 2%Ni, 1¼%Mo–V *steel*

easily shown that from this point of view carbon is harmful because it accentuates the secondary hardening process without improving the tempering resistance at 650°C. This point is illustrated by isothermal tempering curves at 500°C and 650°C shown in Fig.18 for an alloyed 12%Cr steel with carbon contents between 13% and 0·26%. The change in tensile properties with tempering is shown for the same steels in Fig.17. Extremely low proof stress values are obtained in the higher carbon steels after tempering at 550°C while there is little benefit from increased carbon after tempering at 650°C. Nitrogen acts in almost the same way.

From the tempering survey it was shown that two of the most effective elements in increasing tempering resistance are molybdenum and vanadium (Figs.11 and 13). This improvement is most marked in the tempering range 600–700°C, and this increased tempering resistance is also reflected in improved tensile properties after tempering at 650°C. These tensile properties are summarized in Fig.19 from which it can be seen that both Mo and V produce increased tensile and proof stress values and a satisfactorily high PS/TS ratio is maintained. The lower properties obtained with the highest vanadium steels are due to the presence of appreciable amounts of ferrite in the microstructure. For this reason it is concluded that small additions of vanadium (i.e. less than 1%), are most effective.

SELECTION OF PREFERRED COMPOSITIONS

As a result of the work described in this report the effects of C, N, Ni, Mn, Cu, Co, Mo, V, W, Si, Nb, and Ti are known on the austenite/ferrite equilibrium, the A_{c1} and M_s temperatures, the tempering resistance, and the mechanical properties of 12%Cr steels. Consequently, it is possible to select preferred compositions which possess the most desirable combination of properties, namely corrosion resistance, strength, ductility, and impact resistance. Considering the number of alloying elements studied it is perhaps surprising that the preferred composition is a low C and N, 12%Cr–Ni–Mo–V steel and alternatives are only suggested where an improvement is sought in one particular property even at the expense of lowering other properties, i.e. improved strength, together with lower



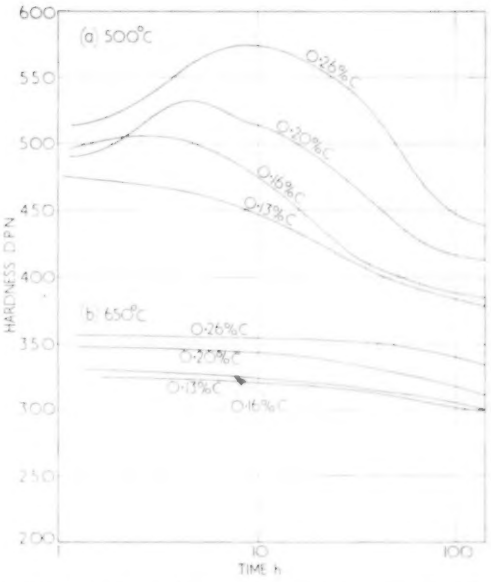
17 *Effect of carbon content and tempering temperature on the mechanical properties of* 12%Cr, 2%Ni, 1½%Mo–V *steel* (b *0·20%* PS, c *0·20%* PS)

impact; or improved ductility, together with lower strength.

First, with regard to carbon and nitrogen, every property is affected adversely if these are high. Adequate strength can be given by the 12%Cr martensitic structure with carbon between 0·08 and 0·15% and nitrogen between 0·01 and 0·04%. With more than these amounts the weldability is reduced owing to high peak hardness in the heat-affected zone. The impact resistance is drastically lowered and, although the hardness and strength are increased, the proof stress values are decreased so that the combination of properties is very much worse.

Secondly, with regard to the alloying elements used, the main objective is to obtain increased tempering resistance. The most effective elements are Mo, W, V, Nb, and Ti and within this group molybdenum and vanadium are preferred. Tungsten has a similar effect to molybdenum but it is more expensive and about twice as much tungsten as molybdenum is required for an equivalent effect. Niobium and titanium have the main disadvantage that they are such powerful carbide-forming elements that they lower the maximum strength which can be obtained unless high solution temperatures are employed to take these carbides into solution and make both the carbon and alloy content fully effective. For most commercial applications this is a serious disadvantage, and consequently the use of these two elements has been avoided. It is intended to publish a separate account of the effect of combinations of Mo, V, Nb, and Ti on the carbide precipitation reactions because these more complex steels have improved properties for some specialized applications, particularly where high-temperature strength is important.

Molybdenum and vanadium have different effects. While molybdenum does improve the tempering resistance its most important effect is to improve the impact properties and for this reason 1·5%Mo is desirable. Above this amount of molybdenum, still further improvement in tempering resistance can be obtained, but since this means that the matrix precipitate is being intensified, the impact properties begin to decrease again. Vanadium has no beneficial effect on impact properties, but a marked effect in increasing tempering resistance. Since it is a strong ferrite-former, however, its use must be limited and it is found that the addition of 0·3%V produces a marked improvement in tensile properties (particularly in the 0·2%PS) without reducing the impact properties too much. The use of larger amounts of vanadium is undesirable since it is difficult to prevent ferrite being formed unless the composition is balanced by the use of a large amount of nickel (which also has disadvantages), and the impact properties are also seriously reduced. A most effective combination, therefore, is 1½%Mo, 0·3%V provided the ferrite-forming effect of these additions is balanced by nickel. Nickel is the preferred austenite-forming element and 2%Ni is adequate. Consequently, the preferred composition is 0·1%C, 12%Cr, 2%Ni, 1½%Mo, 0·3%V. This is air-hardening and has a ferrite-free microstructure at a solution temperature of 1050°C. Because of the low carbon content the ductility and impact properties are good, while the tempering resistance allows high tensile properties to be obtained even after tempering at 650°C. It has been mentioned previously that the undesirable tempering range is 450–550°C, so that the two preferred tempering ranges are 250–350°C and 600–700°C. Typical mechanical properties which can be obtained are:



18 *Effect of carbon on isothermal tempering curves at 500°C and 650°C of a 12%Cr, 2%Ni, 1½%Mo, V steel*

	T 300°C 1 h	T 650°C 1 h
Tensile strength, tons/in²	85–90	60–65
0·2% proof stress, tons/in²	62–67	48–53
Elongation, %	16–20	17–21
Charpy impact, ft/lb at RT	30 min	50 min

This steel represents a considerable improvement compared to the base 12%Cr steel, which has low tensile and impact properties. There may be applications where the improved impact properties are required without such high tensile properties as quoted above. In this case the molybdenum addition would be used without vanadium and a 0·1%C, 12%Cr, 1½%Ni, 1½%Mo composition is obtained. For this steel the equivalent mechanical properties are:

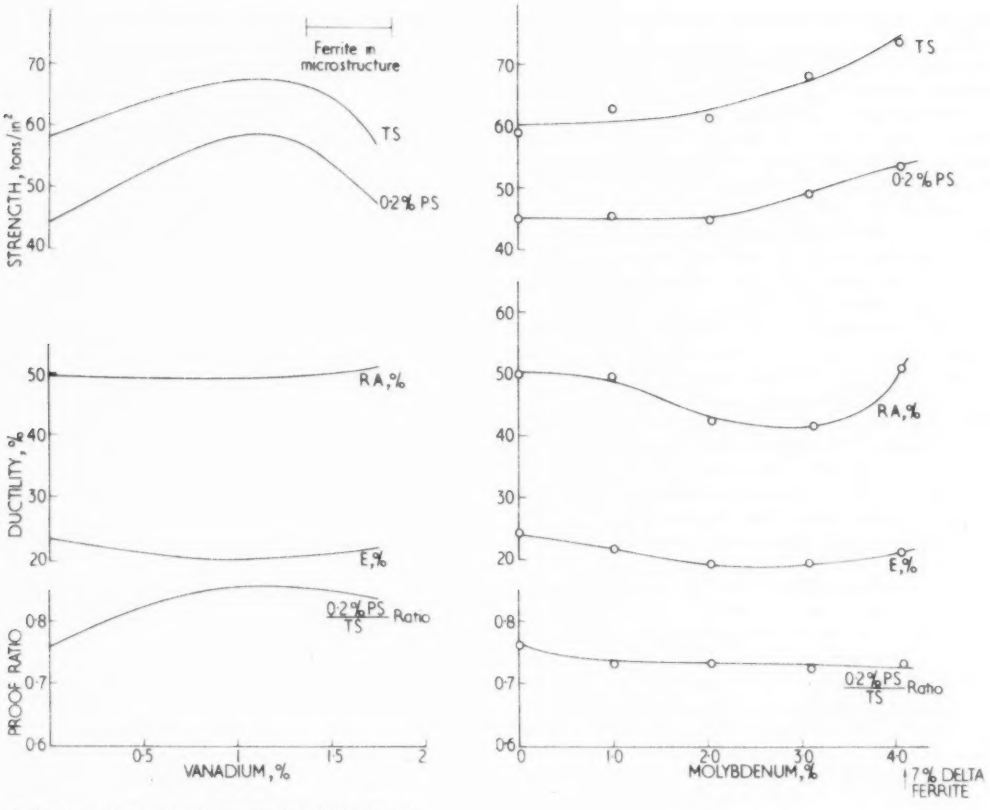
	T 300°C 1 h	T 650°C 1 h
Tensile strength, tons/in²	81–86	50–55
0·2% proof stress, tons/in²	60–65	37–42
Elongation, %	18–22	21–24
Charpy impact, ft/lb at RT	50 min	80 min

It will be seen, therefore, that for general structural applications there are three weldable 12%Cr stainless steels which can provide the different combinations of mechanical properties required. These are:

(i) 0·1%C, 12%Cr
(ii) 0·1%C, 12%Cr, 1½%Ni, 1½%Mo
(iii) 0·1%C, 12%Cr, 2%Ni, 1½%Mo, 0·3%V.

SUMMARY AND CONCLUSIONS

The investigations described were designed to obtain a better understanding of the physical metallurgy of 12%Cr base steels so that the development of compositions required to produce specified properties could be more systematically carried out. The results obtained in this work are briefly summarized below:



solution treated at 1050°C, tempered at 650°C for 1 h

19 *Effect of molybdenum and vanadium on the mechanical properties of a* 0·1%C–12%Cr–2%Ni *steel*

1. The effects of alloying element on the phase equilibrium between austenite and ferrite have been determined mainly for a temperature of 1050°C and, while it is not claimed that the factors for the ferrite-forming tendencies of the elements have any great degree of accuracy, nor even that the effects of the varying elements are linear, the indications provide a basis upon which the constitution of a steel can be assessed and the formation of unacceptably large amounts of delta ferrite avoided.

2. When transformable steels are tempered it is important that they should not re-austenitize at the tempering temperature. Information has been obtained about the effectiveness of varying alloying elements in decreasing the A_{c1} temperature and of these, Ni and Mn are most effective so that if re-austenitization is not to occur below 700°C a limit of 2%Ni must be imposed upon the steel. On the other hand, Si, Al, Mo, and especially V, are all effective in raising the A_{c1} temperature, in which case it is possible that more Ni can be accommodated.

3. The M_s temperature is important because it is desirable that no austenite should be retained in the solution-treated condition, and information about the depression of the M_s as a function of the alloying element concentration has been taken mainly from the work of Steven and Haynes.[2] From these results it is possible to design steels in which the M_s temperature is not too low.

4. The transformation characteristics of 12%Cr steels show that not only is the pearlite reaction preceded by carbide precipitation, but also that the pearlite reaction is extremely slow and does not go to completion within practicable periods of time, i.e. 24 h. Also, the bainite reaction is completely suppressed. It is extremely difficult, therefore, to soften alloyed 12%Cr steels by isothermal transformation techniques. It is best to temper the martensitic steel at the highest temperature which does not produce re-austenitization (usually about 700°C) but this is still difficult in the alloyed steels which have considerably improved tempering resistance.

5. The effects of varying alloying elements on the tempering characteristics of alloyed 12%Cr steels have been determined and related to the structural changes which take place. It has been shown that secondary hardening can occur at temperatures within the range 400–500°C with the precipitation of a Cr-rich carbide of the M_2X type. Various alloying elements have been shown to stabilize this carbide, thereby increasing the temperature range over which it can form, and also increasing the coherency strain associated with it so that the secondary hardening is increased. Overageing is accompanied by the formation of $M_{23}C_6$ at grain boundaries and a gradual re-solution of the M_2X. It has been shown that carbon and nitrogen are very effective in accentuating the secondary hardening of 12%Cr steels but that Ni and Mn are much less

effective, and in fact, can cause some acceleration in the rate of softening of these steels. Copper produces a marked secondary hardening reaction which is mainly due to an intensification of the M_2X precipitate but could also be the result of the formation of a Cu-rich phase which has not been identified. The main effect of Co is to cause marked solid solution hardening.

The ferrite-forming elements Mo, W, V, and Si have very similar effects in that they not only cause some solid solution hardening, but also intensify the secondary hardening reaction. V also slightly intensifies the secondary hardening reaction but its main effect is in the stabilization of M_2X so that higher hardnesses are produced at high tempering temperatures. Si causes a slight secondary hardening owing to its effect upon the ferrite lattice parameter and also solid solution hardens to a small extent at high temperatures. The effects of Ti and Nb are complicated because two factors must be taken into consideration, namely:

(i) the effect of Ti and Nb in removing carbon from solid solution, thereby decreasing the hardness or strength at any tempering temperature

(ii) the formation of delta ferrite.

It has been shown, however, that both Ti and Nb can accentuate the secondary hardening effect and also stabilize the M_2X to higher temperatures so that they can impart considerable tempering resistance, but unfortunately it is often necessary to use high solution treatment temperatures in order to obtain sufficient of the element in an effective form.

6. In considering the mechanical properties of 12%Cr steels it has been shown that secondary hardening is accompanied by a decrease in the proof ratio and very poor impact properties. This is the result of coherency straining due to M_2X formation. Up to 2%Mo improves both the strength and the impact resistance. Above 2%, however, its effect upon the M_2X precipitate outweighs its beneficial effect on impact properties. Vanadium has a beneficial effect upon the tensile strength and can be used in small amounts without too adverse an effect on the impact properties. Increasing carbon and nitrogen intensifies the secondary hardening reaction and leads to very much worse proof ratios and impact properties. For the best combination of strength and impact a 0·1%C, 12%Cr, 2%Ni, 1½%Mo, 0·3%V composition is preferred. Where better impact and ductility are required and a reduction in tensile strength can be accepted, then a 0·1%C, 12%Cr, 1½%Ni, 1½%Mo composition can be used.

7. It has been shown that there are two temperature ranges in which it is advantageous to temper alloyed 12%Cr steels, namely between 250° and 350°C and 600° and 700°C; in these two temperature ranges very attractive combinations of mechanical properties can be obtained at different strength levels. Tempering in the temperature range between 450° and 500°C is most undesirable as the properties are very poor.

ACKNOWLEDGMENTS

The authors wish to thank Mr F. H. Saniter, Director of Research, The United Steel Companies Ltd, for permission to publish this paper.

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Under-hearth cooling for blast-furnaces

E. M. Summers

INTRODUCTION

THE PREVENTION of erosion of the hearth brickwork of blast-furnaces is a problem with which all plants have to contend, and the extent of the erosion can vary according to the type of iron produced.

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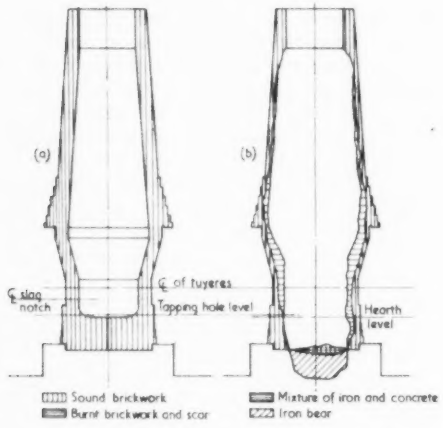
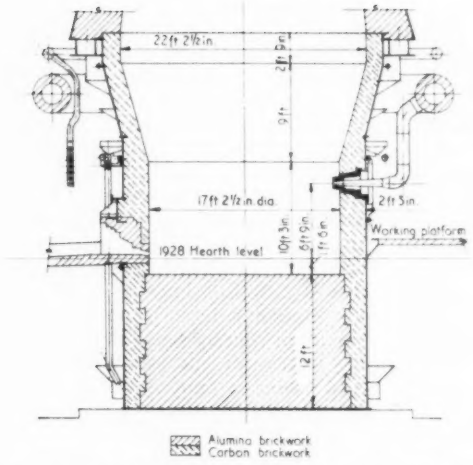
SYNOPSIS

The problem of blast-furnace hearth erosion is considered and theoretical calculations on the effect of cooled carbon brickwork are given. The design of two furnaces employing under-hearth cooling is described and the operating data for one installation is given in detail. Finally, conclusions are drawn on the use of similar constructions as a general practice in blast-furnace design. 1846



(a) blown-in March 1949 (b) blown-out March 1956

1 *Penetration of iron into foundation concrete of no.1 furnace*



2 *Modification of nos.1 and 2 furnaces by adding carbon side walls and spray cooling the hearth jacket*

The problem appears to have received more active consideration in other countries, and furnaces are known to be working in Austria, Sweden, and the USSR with some form of cooling under the hearth. In general the construction consists of a carbon pad laid on firebrick with the cooling below the firebrick.

At Stanton the problem of hearth erosion has been particularly acute when operating on high-temperature foundry irons, and iron has been found to have penetrated through the firebrick pad well down into the foundation concrete, as illustrated in Fig.1.

The Stanton furnaces as originally designed had firebrick pads 7 ft 6 in. thick, and in the case of nos.1 and 2 furnaces, which stood on concrete plinths, it was possible to remove the concrete, extend the hearth jacket, and increase the depth of the hearth pad. By putting in carbon side walls and spray cooling the hearth jacket, conditions on these two furnaces were greatly improved, and this construction is shown in Fig.2. In the case of nos.3, 4, and 5 furnaces, however, the top of the concrete plinth was at ground level and did not permit any extension of the hearth jacket without major reconstruction of the furnaces, and consequently it was not possible to increase the thickness of the hearth pad. Even with carbon side walls it was not considered that the thickness of the firebrick pad was sufficient to contain the downward penetration of the iron, and consideration was therefore given to the installation of an under-cooled carbon hearth.

Theoretical considerations

Calculations were therefore made for a total pad depth of 6 ft 3 in. with varying thicknesses of firebrick below the carbon in order to study the depth of the freezing line and interface temperature. It has been concluded from researches carried out by BISRA that the thermal conductivity of carbon above 1000°C is 3·4 Btu/ft h°F and the hot metal temperature was taken as 1500°C. Details of these calculations are given in Table I.

It was obvious that the effect of putting firebrick under the carbon was to reduce the benefit gained from the high conductivity of the carbon and consequently to lower the freezing line. Therefore it was considered that the firebrick should be excluded and a design was prepared in which the carbon was placed directly on the cooled bottom plate. With this design the theoretical indication was that iron penetration would be arrested at a depth of 1 ft 3 in. below the hearth level.

Voice* had already suggested that air might with advantage be substituted for water as the cooling medium and it was certainly considered to be safer in the event of an unexpected penetration through the hearth bottom. It was therefore decided to use air as the cooling medium.

BISRA were asked to make new calculations for the revised construction and a summary of the methods used is presented.

The heat flow down through the hearth can be considered in two parts, that directly beneath the hearth and that below the walls. That through the centre cylinder may be considered as uniform over the cross-section. The outer annulus, however, necessitated using the BISRA analogue to estimate the temperature distribution and heat flow. As before, thermal conductivity of carbon was taken as 3·4 Btu/ft h°F and the hot metal temperature at 1500°C. Thus the heat flow through the hearth was estimated as

$$\frac{3{\cdot}4}{6{\cdot}33} \times 2350 = 1270 \text{ Btu/ft}^2\text{h}$$

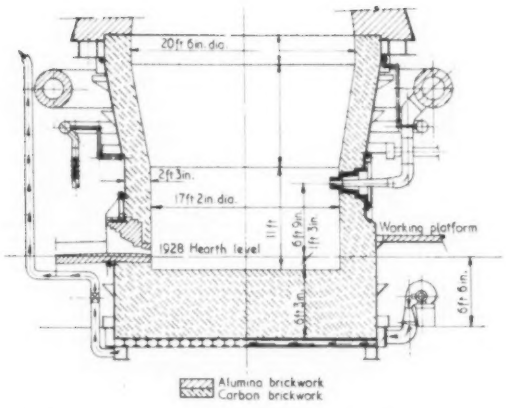
and the heat flow through the annulus was estimated as

540 Btu/ft²h

* E. W. Voice: *Iron Steel Eng.*, 1952, **29**, July, 125–128.

TABLE I Details of calculations for pad depth of 6 ft 3 in.

Thickness of firebrick, in.	Depth of freezing line, in.	Heat flow down through hearth, Btu/ft²h	Heat flow down through base, Btu/ft²h	Total vertical heat flow, Btu/h	Interface temperature, °C
0	15	1500	730	466000	30
6	21	1100	515	323000	512
12	27	875	416	258000	784
18	32½	720	355	216000	970
24	38½	600	300	179000	1100

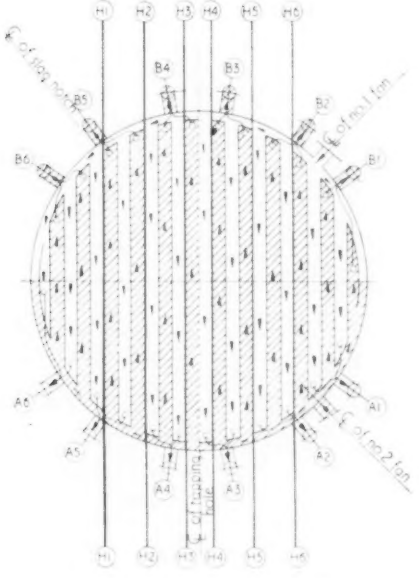


3 *Design for under-cooled carbon hearth, no.3 furnace*

Therefore the necessary heat extraction rate was estimated to be 412240 Btu/h giving an average heat flow of 930 Btu/ft²h.

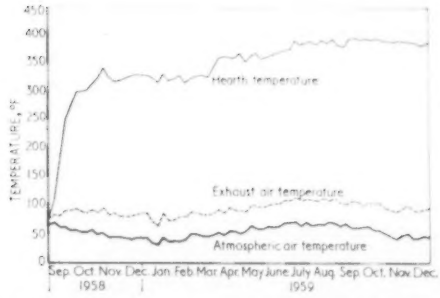
The rise in temperature of the air passing beneath the furnace depends upon the heat transfer co-efficient from the duct wall to the air. This may be calculated from the Reynolds number for the air flow. Since it was proposed to use standard fans it was known that the air flow through the centre duct would be about 500 ft³/min. although the calculation of the heat transfer co-efficient was relatively insensitive to this value. The Reynolds number proved to be 77000. The Prandle number for air is 0·74. Thus the heat transfer co-efficient is

$$(60)\,(0{\cdot}023)\,(0{\cdot}000307)\,(0{\cdot}74)^{0{\cdot}4}\,(77000)^{0{\cdot}8} = 3{\cdot}0 \text{ Btu/ft}^2\text{h}^\circ\text{F}$$



A slide valves and thermocouples in exhaust air duct from fan 1
B slide valves and thermocouples in exhaust air duct from fan 2
H thermocouples in tubes set between the mild-steel baseplate and hearth bricks

4 *Air flow arrangement for no.3 furnace*



5 *Weekly average hearth temperature, exhaust air temperature, and atmospheric air temperature of no.3 furnace in first 16 months' operation*

Thus the temperature difference between the air and duct wall would be

$$\frac{930}{3{\cdot}0} = 310^\circ\text{F}$$

If the inner surface of the duct has an average temperature of 380°F then the effective mean temperature of the air flowing through the duct would be

$$380 - 310 = 70^\circ\text{F}$$

The average air temperature is 50°F, so the logarithmic mean temperature difference formula gives an exit temperature of about 90°F. Thus the temperature rise would be 40°F.

In order to extract the necessary 412240 Btu/h of heat the volume of air required would be

$$= \frac{412240}{(0{\cdot}0194)\,(40)\,(60)} = 8850 \text{ ft}^3/\text{min}$$

With a given airflow, then as the inlet air temperature varies so the exit air temperatures will also vary in sympathy giving a constant difference of about 40°F.

Details of construction for no.3 furnace

After consideration of these theoretical factors and practical site conditions it was decided to install an under-cooled carbon hearth using air as the cooling medium, and the hearth design is shown in Fig.3.

Mild-steel joists 6 in. × 5 in. were laid parallel below the hearth and the ducts completed by welding plates across the joists thus making an air- and water-tight bottom. Two fans were provided each capable of delivering 4400 ft³/min of air at a pressure drop of 7 in. wg and a connexion was made to the cold-blast main for emergency use. The air flow arrangement is shown in Fig.4 and it will be seen that alternate ducts were connected together and the exit flues provided with dampers so that the flow through any section could be controlled. The hearth walls were constructed of standard carbon bricks and the pad of interlocking corrugated blocks 12 in. × 12 in. × 20 in. Six thermocouples were placed on the top of the joists, and after levelling with carbon ramming the carbon blocks were placed in position. The carbon side walls were cooled with external sprays on the hearth jacket.

Operating data for no.3 furnace

Although the design work was carried out in 1955 and 1956, and construction in 1957, owing to trade conditions the furnace was not blown in until August 1958.

TABLE II Alterations in slide settings from 28 August 1958

Date	Setting, in. A1	A2	A3	A4	A5	A6	B1	B2	B3	B4	B5	B6	
28.8.58	1	2	3	3	2	2	1	2	3	3	2	2	Fans started at 11.45 am
4.9.58	2	3	4	4	3	3	2	3	4	4	3	3	All opened 1 in. at 9.45 am
4.9.58	2	3	5	4	3	3	2	3	5	4	3	3	A3 and B3 opened 1 in. at 4.10 am
5.9.58	2	3	6	5	3	3	2	3	6	5	3	3	A3, B3, A4, and B4 opened 1 in. at 3 pm
6.9.58	2	3	7	6	3	3	2	3	7	6	3	3	A3, B3, A4, and B4 opened 1 in. at 4 pm
7.9.58	2	3	7	7	3	3	2	3	7	7	3	3	A4 and B4 opened 1 in. at 9.45 am
9.9.58	2	3	8	8	3	3	2	3	8	8	4	3	A3, B3, A4, and B4 opened 1 in. at 11 am
10.9.58	2	3	10	10	3	3	2	3	10	10	4	3	A3, B3, A4, and B4 opened 2 in. at 9.30 am
11.9.58	2	3	10	10	2	2	2	3	10	10	3	2	A5, B5, A6, and B6 closed 1 in. at 2.50 pm
15.9.58	1	3	10	10	2	1	1	3	10	10	3	1	A1, B1, A6, and B6 closed 1 in. at 2.10 pm
18.9.58	1	3	10	10	2	1	1	3	10	10	4	1	B5 opened 1 in. at 2.55 pm

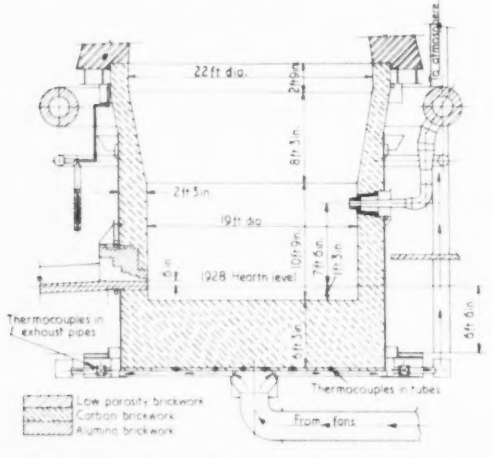
Note: A slide is fully open at 10 in.

Figure 5 shows in graphical form details of the temperatures over the first 16 months' operation. Table II shows the alterations made in the various slide settings in the exit flues to achieve a uniform temperature over the whole of the hearth bottom. This proved possible although, as was expected, most of the air had to be forced across the centre of the furnace and very little under the side walls.

Although Fig.5 shows the weekly average hearth temperature of the six thermocouples below the carbon, in fact there has been very little variation in the individual figures, and the same applies to the average discharge temperatures from the flues.

For protection during the blowing-in period 9 in. of firebrick was put over the top of the carbon and it is assumed that this disappeared some time between the beginning of September and the beginning of November 1958 which is the period of greatest temperature gradient.

From then on the rise in temperature has been very gradual and has been partly related to an increase in air temperature due to summer conditions. The whole of the increase cannot however be attributed to changes in air temperature and it is considered that it is due partly to erosion and partly to an increase in the conductivity of the top course of carbon due to impregnation with iron and to graphitization. The increase in temperature, however, after the initial blowing-in rise is only 60°F which does not indicate any serious erosion, and all the indications are that the carbon is almost completely intact.



6 *Design of no.5 furnace to give maximum cooling at hearth centre*

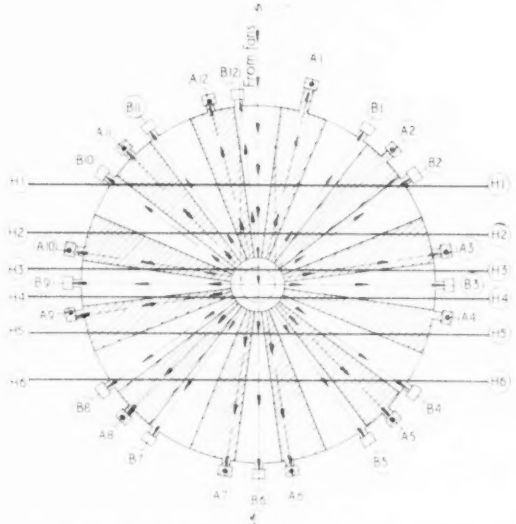
The furnace has operated well, there has been no difficulty in maintaining hearth temperature, and the coke rate has been no higher than on furnaces of normal firebrick construction. The emergency cold air connexion has only been used once when it became necessary to change one of the fans.

Details of design for no.5 furnace

Experience with no.3 furnace has suggested that the design would be improved if the maximum cooling could be obtained at the centre of the hearth. Figures 6 and 7 show the design and air flow arrangement for no.5 furnace. The hearth construction remains as before but cold air is taken direct to the centre of the hearth and comes out along a series of radial ducts, and this arrangement gives the maximum of cooling at the point where it is most needed. The capacity of the fans has also been increased to provide a reserve against any increase in conductivity which may arise due to iron impregnation or erosion.

SUMMARY AND CONCLUSIONS

Theoretical considerations have indicated the advantages to be gained from a cooled carbon pad in



A slide valves only in exhaust air duct
B slide valves and thermocouples in exhaust air duct
H thermocouples in tubes set between the mild-steel baseplate and hearth bricks

7 *Air flow arrangement for no.5 furnace*

preventing hearth erosion. They have also shown that a satisfactory construction can be evolved without a firebrick insulating course below the carbon, and that air is a satisfactory coolant.

Details of the design of Stanton no.3 furnace hearth and a modified design for Stanton no.5 furnace hearth are described based on the theoretical considerations. The operating data for the first 16 months of the campaign on no.3 furnace is given and this indicates that hearth penetration has been almost completely arrested without adverse effect on blast-furnace operation.

It is considered that while these two installations were designed to meet a special problem at Stanton, the principles may have a universal application in the construction of blast-furnace hearths. On a new installation the cost of under-hearth cooling using a carbon pad is not likely to be much more than a conventional construction employing a thick firebrick pad and a deep hearth jacket. The advantages to be gained from this type of construction in preventing hearth penetration and consequent hearth breakouts are considerable, and if, as is expected, the formation of a salamander has been prevented, there should be substantial savings in relining costs to offset against any additional construction cost.

ACKNOWLEDGMENTS

The author wishes to thank Mr E. W. Voice and the staff of the Ironmaking Division of BISRA for help and advice, particularly in respect of the calculations. Also the staff of the Stanton blast-furnace department and drawing office who were responsible for the detailed design, installation, and operation of the under-hearth cooled furnaces.

Effect of Al and N on the occurrence of intergranular fracture in steel castings

B. C. Woodfine, Ph.D., B.Met. and A. G. Quarrell, D.Sc., Ph.D., A.R.C.S., D.I.C.

SYNOPSIS

An investigation has been made into the occurrence of intergranular fracture in low alloy steel castings of varying aluminium and nitrogen contents, and it has been confirmed that this type of fracture is associated with the precipitation or segregation of aluminium nitride at the austenite grain boundaries on slow cooling below about 1150°C. An explanation is offered for the absence of intergranular fracture from steels containing very large amounts of aluminium nitride and it is shown that intergranular fracture can be avoided by deoxidizing with titanium, zirconium, or titanium in combination with aluminium. The effect of heat-treatment on the occurrence of intergranular fracture has also been examined. 1825

INTRODUCTION

A peculiar form of fracture characterized by large smooth facets and having the general appearance shown in Fig.1 is occasionally found in carbon and low alloy steel castings. It is associated with a low tensile ductility and low impact values, and is usually referred to as 'intergranular' or 'rock-candy' fracture. As part of an investigation into the occurrence of this form of fracture in commercial castings it was decided to re-examine the conclusion reached by Lorig and Elsea,[1] that aluminium nitride precipitation at the primary austenite grain boundaries was one of the principal causes of intergranular fracture. Although the term intergranular fracture will be used without qualification, it is apparent from the size of the facets that this term is not strictly accurate. The fractures do not occur around the grain boundaries present in the steel at the temperature of testing but around what were either the delta or the austenite grain boundaries.

EXPERIMENTAL

The experimental steels were made in high frequency furnaces using base materials of different nitrogen contents (*see* Table I). After the alloying additions had been made to bring the melt to the required composition, the metal was cast into green sand moulds to give either keel block or simple cylindrical rod castings.

Preliminary experiments on a commercial steel showing intergranular fracture demonstrated that the presence of the intergranular weakness could be masked by brittle cleavage fracture if the specimens were broken in the annealed or normalized condition, so the sections cut from the castings were oil-quenched from 900°C, tempered at 650°C for 1 h, and then water-quenched to avoid any possible occurrence of temper brittleness. After heat-treatment the specimens were notched with a saw cut and broken by slow bending in a hydraulic testing machine. The extent of intergranular fracture was determined by inspection of the fractured surfaces.

The total nitrogen content of a steel was determined by dissolving a sample in dilute sulphuric acid, fuming for 1 h, distilling with caustic soda solution, absorbing the ammonia released in boric acid solution, and finishing absorptiometrically after colouration with Nessler's reagent. The acid-soluble nitrogen was determined by the same method, omitting the fuming treatment. The aluminium nitride content of the steel was estimated by the ester halogen separation method

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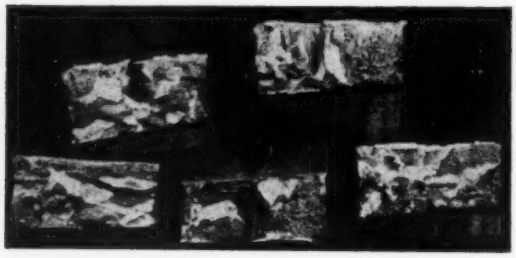
Professor Quarrell is at the Department of Metallurgy, University of Sheffield. Dr Woodfine, formerly of the Department of Metallurgy, University of Sheffield, is now at the GEC Atomic Energy Division, Erith, Kent.



1 *Typical example of intergranular fracture in heat-treated samples from a steel casting*

devised by Beeghly.[2–4] The acid-soluble aluminium content was determined by the standard phosphate method.[5]

PRELIMINARY EXPERIMENTS

The first series of 15 lb keel block castings were made using the side-blown converter scrap adjusted to the rough composition 0·3%C, 0·5%Si, 1%Mn, 0·5%Cr, 1%Ni, with aluminium additions of 0, 7, 12, 19, 28, and 37 lb/ton, and were allowed to cool to room temperature in the moulds. Examination of the fractures of the hardened and tempered specimens showed that the only casting to contain any trace of intergranular fracture was that treated with an aluminium addition of 7 lb/ton.

The next series of casts were made in the same way to the same basic composition with aluminium additions of 0, 5, and 9 lb/ton but were knocked out from the moulds as soon as they had solidified and transferred to a furnace at 980°C. After 1 h at this temperature they were cooled over 20 h to 300°C in the furnace and then air-cooled to room temperature. The casting made with no aluminium addition showed no intergranular fracture but the fractures of both the aluminium-treated castings had 30–40% intergranular fracture (Fig.2). Metallographic examination of the two aluminium-treated castings revealed a fine precipitate following what were presumably the prior austenite grain boundaries (Fig.3). The precipitate linked up with the intergranular fracture facets and appeared identical to that observed by Lorig and Elsea[1] and identified by them as aluminium nitride. No intergranular precipitates could be detected in the first series of more rapidly cooled castings.

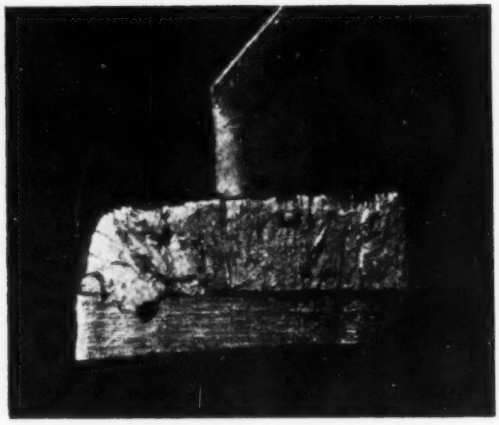
These preliminary experiments demonstrated that a form of intergranular fracture could be produced in low alloy steel castings by treatment with aluminium, provided that a sufficiently slow cooling rate was achieved after solidification, and that this intergranular fracture was associated with a grain boundary precipitate of aluminium nitride.

EFFECT OF ALUMINIUM ADDITIONS ON STEELS OF VARIOUS NITROGEN CONTENT

In this series of melts, electrolytic iron, a plain carbon steel, and Armco iron were used as base materials to

TABLE I Nitrogen content of the base materials used

Base material	Acid-soluble nitrogen, %
Side-blown converter scrap	0·015
Electrolytic iron	>0·02
High purity plain carbon steel	0·0105
Armco iron	0·005

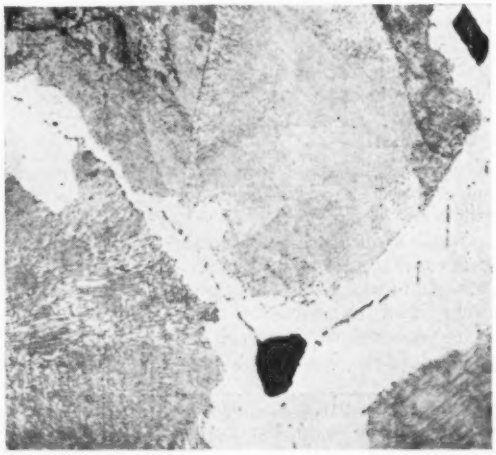


2 *Intergranular fracture in heat-treated* Al-*killed casting*

give high, medium, and low nitrogen contents respectively in the final castings. Additions were made to the melts to give compositions of about En100 after which a number of 1 in. dia. rods with simple cylindrical feeder heads were sand-cast, each with a different aluminium addition. The castings were knocked out of the moulds immediately after solidification and transferred to a furnace. The low and medium nitrogen castings were held at 950°C for 3 h, at 900°C for a further 3 h and then furnace cooled to room temperature. The high nitrogen castings were held at 980°C for 2 h, cooled to 900°C for a further 3 h, and then furnace cooled.

Fracture tests were made on a heat-treated sample from each casting with the results given in Table II. The results show that with an acid-soluble nitrogen content of about 0·005%, 100% intergranular fracture can be obtained with high aluminium additions (7 or 14 lb/ton), and that with a high total nitrogen content (about 0·015%) relatively low aluminium additions can also produce 100% intergranular fracture.

The results for the high nitrogen castings (B1–B6)



3 *Intergranular precipitate in* Al-*killed casting, as-cast condition, etched in picral*

TABLE II Effect of Al on the occurrence of intergranular fracture in steels of varying nitrogen content

Cast	Deoxidants added	Base material	Nitrogen content Acid soluble, %	Ester halogen, %	Total, %	Acid-soluble Al, %	Intergranular fracture in the hardened and fully tempered condition, %
B1	Fe–Si only	Electrolytic iron	0·0031	0·0017	0·0148	0·01	5
B2	1½ lb/ton Al		0·0065	0·0051	0·0163	0·06	80
B3	3½ lb/ton Al		0·0173	0·0140	0·0173	0·142	95
B4	7 lb/ton Al		0·0143	0·0116	0·0153	0·230	100
B5	14 lb/ton Al		0·0159	0·0154	0·0165	0·366	100
B6	28 lb/ton Al		0·0162	0·0165	0·0220	0·865	0
C1	Fe–Si only	High purity	0·0070			0·012	0
C2	1½ lb/ton Al	plain carbon steel	0·0107			0·017	0
C3	3½ lb/ton Al		0·0106			0·158	60
D1	Fe–Si only	Armco iron	0·0048				0
D2	7 lb/ton Al		0·0054				100
D3	14 lb/ton Al		0·0046				100

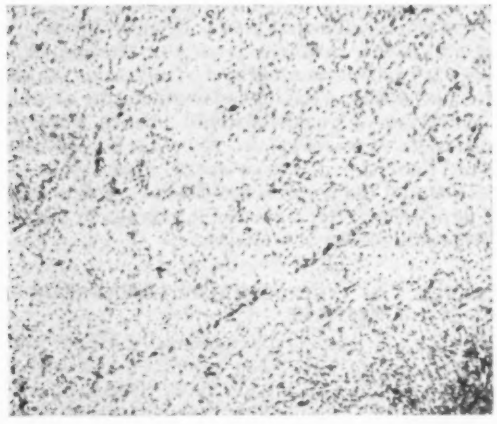
have two noteworthy features: (*a*) although no deliberate aluminium addition was made to the first casting and this only had acid-soluble and ester halogen nitrogen contents of 0·0031 and 0·0017% respectively, there was a trace of intergranular fracture present, and (*b*) the amount of intergranular fracture increased with increasing aluminium additions up to 14 lb/ton but with an aluminium addition of 28 lb/ton, intergranular fracture was entirely absent.

Metallographic examination of specimens cut from the heat-treated pieces of the high nitrogen castings showed that there was no visible grain boundary precipitate in cast B1, and that in cast B2 only very faint traces of such a precipitate could be found, although the amount of intergranular fracture observed was about 80%. In casts B3, B4, and B5 there were obvious grain boundary precipitates present (Figs.4–6); the precipitate becoming coarser as the aluminium content of the steel was increased. In cast B5 the precipitate consisted almost entirely of needles which had grown away from what was presumably the original austenite boundary into the grains on either side. Examination under polarized light showed that the needles were anisotropic, and their general appearance was identical with the precipitate identified as aluminium nitride by Lorig and Elsea. In cast B6, which showed no intergranular fracture although it had an extremely high ester halogen nitrogen content, 0·0165%, there was no sign of a grain boundary precipitate but there were many large, dark grey, angular particles of a type not previously observed (Fig.7). Many of the particles were hexagonal in shape and their general appearance and strong anisotropic response to polarized light agreed with the published data for aluminium nitride.[6,7]

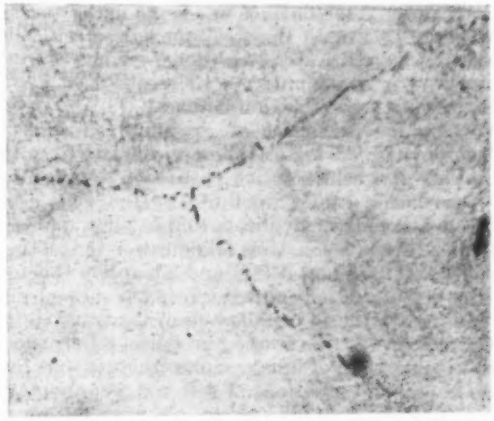
EFFECTS OF TITANIUM AND ZIRCONIUM ADDITIONS

The results described above demonstrate that aluminium nitride can produce a form of intergranular fracture in cast steels, and since both titanium and zirconium form nitrides which are more stable than aluminium nitride, a few casts were made to see what was the effect of adding these elements, with or without aluminium.

To examine the effect of titanium additions three 15 lb keel block castings were made from the side-blown converter-scrap adjusted to the composition of about 0·3%C, 0·5%Si, 1%Mn, 0·5%Cr, 1%Ni, with additions of ferrotitanium 19 lb/ton and aluminium 5 lb/ton, separately and together. After casting, the keel blocks were knocked from the moulds and transferred to a furnace at 980°C for 1 h before cooling in the furnace to room temperature. Examination of the fractures of the heat-treated slices from the blocks showed that intergranular fracture was only present in the casting to which aluminium by itself had been added. The analytical results obtained on the castings and the mechanical properties of the heat-treated slices are given in Tables III and IV respectively. It is apparent that even in the presence of 0·18%Al and with a total nitrogen content of 0·014%, 0·08%Ti prevents the occurrence of intergranular fracture without adversely affecting the mechanical properties of the



4 *Grain boundary precipitate in cast B3 in quenched and tempered condition, etched in picral*



5 *Grain boundary precipitate in cast B4 in quenched and tempered condition, etched in picral*

6 *Grain boundary precipitate in cast B5 in quenched and tempered condition, etched in picral*

7 *Inclusions of* MnO–FeO, Al_2O_3, *and* AlN (*hexagonal particles*) *in cast B6*

steel. The analytical results show that in the steel containing titanium, the ester halogen nitrogen is considerably greater than the acid-soluble nitrogen, which agrees with the observations of Beeghly.[2] This is presumably because some of the Ti (C, N) compounds were caused to break down by the bromine used in the ester halogen separation. It will be noted that the acid-soluble nitrogen content of the steel deoxidized with aluminium plus titanium is less than that of the steel treated with titanium alone (*cf.* A18 and A16 of Table III). Efficient deoxidation by aluminium probably leaves most of the titanium to combine with nitrogen. The relative stabilities of Al_2O_3 and TiO_2 would accord with this view.

In the steels containing titanium, in addition to inclusions of TiC, TiN, and MnS, a tan-coloured phase was also observed with the same general appearance as the sulphides observed by Sims *et al.*[8] in steels deoxidized with titanium. This phase was probably a Ti–Mn–Fe sulphide and although present as grain boundary films and in the form generally referred to as Type II sulphide inclusions[8, 9] its presence did not result in a decreased ductility nor in the occurrence of intergranular fracture. No grain boundary precipitate other than the sulphide films was found in the castings containing Ti, but in the steel deoxidized with aluminium alone there were occasional strings of grain boundary particles similar to those observed in the other casts treated with aluminium.

To examine the effect of zirconium additions a series of cylindrical rod castings were made from the plain carbon steel scrap, to the rough composition 0·6%C, 2%Mn, 0·1%Si, 0·6%Cr, 0·8%Ni using different amounts of a silicon–40%Zr alloy and aluminium as deoxidizers. After solidification the castings were treated for 3 h at 950°C, and 3 h at 900°C before furnace cooling. The fractured specimens showed that an addition of the Si–Zr alloy equivalent to 2 lb/ton to the melt just before pouring produced a fully-killed steel with no trace of intergranular fracture, but that when aluminium additions of 4 lb/ton were made to the mould, Si–Zr additions to the melt of up to 4 lb/ton were completely ineffective in preventing the occurrence of intergranular fracture.

EFFECT OF HEAT-TREATMENT ON THE OCCURRENCE OF INTERGRANULAR FRACTURE

Beeghly[3] has shown that aluminium nitride can be taken into solution in austenite by heating in the range 1100–1250°C, and retained in solution by quenching. It was decided therefore to examine the effect of heat-treatment on the occurrence of intergranular fracture, and to this end a number of experiments were made on cast B2 which showed 80% intergranular fracture in the heat-treated condition. Specimens were oil-quenched and tempered at 650°C after heating in an atmosphere of argon at temperatures up to 1250°C for various times. Although the amount of intergranular fracture was apparently reduced by heating at temperatures of 1100 and 1150°C, it still persisted after treatment at 1200°C for 5 h. Complete removal could only be achieved by heating at 1250°C for times of more than about 2 h. To see whether intergranular fracture would reappear on holding at lower temperatures after the high temperature treatment, specimens were heated at 1250°C for 2 h, and were then cooled in the furnace to 900°C and held for 20 h before being oil-quenched and tempered. The fractures of these specimens were found to be 100% intergranular in nature, but they had a different macroscopic appearance from those in the original heat-treated casting, being composed of small polyhedral facets. The identity of these facets with the austenite interfaces established by heating at 1250°C was confirmed by sectioning the fractures. To determine the highest temperature at which the intergranular weakness would develop, a number of specimens were heated at 1250°C for 2 h, cooled to various lower temperatures for 1 h and then oil-quenched and tempered before fracturing. Specimens held at or below 1100°C showed intergranular fracture at the new austenite boundaries.

OCCURRENCE OF INTERGRANULAR FRACTURE IN COMMERCIAL CASTINGS

A number of commercial castings showing intergranular fracture have been examined and without attempting to discuss these in detail there are several points which merit comment. Two cases were found where the fracture was almost certainly not due to aluminium nitride alone. In the first example, a C–Mn

TABLE III Nitrogen content of Ti and Al deoxidized steels

Cast	Deoxidant used	Nitrogen content Acid soluble, %	Ester halogen, %	Total, %	Acid, soluble Al, %	Ti, %
A16	Fe–Ti	0·0026	0·0056	0·0137	0·06	0·15
A17	Al	0·0093	...	0·0119	0·14	Nil
A18	Fe–Ti+Al	0·0018	0·0057	0·0137	0·18	0·08

steel wheel-centre casting, intergranular fracture was associated with grain boundary films of ferrite, and was not present when the specimens were quenched and tempered. In the second case, an ingot of a 4%Ni–Cr steel, the intergranular weakness was completely removed at 1250°C and could not be made to reappear by holding at lower temperatures or by slow cooling. All of the castings, however, had ester halogen nitrogen contents greater than the lowest value at which intergranular fracture was observed in the experimental casts, and, except for the two cases mentioned above, there was some correlation between the ester halogen nitrogen content and the extent of intergranular fracture. A number of high temperature heat-treatments were carried out on samples from a 6 in. thick armour-plate test casting and it was found that it showed almost exactly the same response to heat-treatment as did the experimental cast B2. The intergranular fracture was still present after heating at 1200°C for 5 h but could be removed at 1250°C by heating for more than 1½ h. Specimens given this heat-treatment and subsequently cooled to lower temperatures showed intergranular fracture at the new austenite grain boundaries after holding for 1 h at or below 1140°C. Specimens held at 900°C for various times were found to give about 10% intergranular fracture after ¼ h, 50% after 1 h, and 100% after 20 h. This particular steel had an acid-soluble nitrogen content of about 0·009%, an ester halogen nitrogen content of 0·008%, and an aluminium content of 0·07%.

DISCUSSION

The results described above show that aluminium nitride can produce a form of intergranular fracture in low alloy steel castings if these are slowly cooled after solidification or are held in the temperature range 800–1100°C, and that the intergranular fracture is most evident when the steel is in the quenched and tempered condition. The heat-treatment results also show that, in agreement with the work of Beeghly[3] and Leslie *et al.*[10] on the solubility of aluminium nitride in austenite, intergranular fracture can be removed by heating at 1250°C but reappears on cooling below about 1150°C, although it then occurs at the interfaces of the austenite grains established by the high temperature heat-treatment. Although in many cases of intergranular fracture a grain boundary precipitate of aluminium nitride in the form of particles, rods, or needles could be detected metallographically, this was not always so; in one steel showing 80% intergranular fracture there were only very faint traces of a precipitate. All these observations are in agreement with the work of Lorig and Elsea.[1]

The results obtained with the high nitrogen base material (series B), where it was observed that intergranular fracture was not present with an aluminium addition of 28 lb/ton, can be explained by reference to the metallographic observations. With ester halogen

TABLE IV Mechanical properties of Ti and Al deoxidized steels

Cast	Deoxidant used	Mechanical properties of the hardened and tempered steel Yield point, tons/in²	TS, tons/in²	Reduction of area, %	Fracture
A16	19 lb/ton Fe–Ti	44·1	51·5	18	Ductile
A17	5 lb/ton Al	41·6	50·0	4·9	Intergranular
A18	5 lb/ton Al and 19 lb/ton Fe–Ti	43·3	50·8	21·5	Ductile

nitrogen contents of up to 0·005% it is doubtful whether any grain boundary precipitate is visible, but on increasing the ester halogen nitrogen content to between 0·0116 and 0·0154% continuous chains of precipitated particles occur, changing in appearance from rounded particles and small needles to large needles growing out from the grain boundary. The precipitate becomes coarser with increasing aluminium content until at about 0·8%Al and an ester halogen nitrogen content of 0·0165% the precipitate ceases to occur at the grain boundaries at all, and is replaced by general precipitation within the grains of more or less uniformly distributed angular shaped particles of aluminium nitride. This change is associated with the complete disappearance of intergranular fracture. The size and distribution of the particles suggest that in casts B2–B5 precipitation of aluminium nitride occurs after solidification, whereas in cast B6 aluminium nitride nuclei exist in the melt and any further precipitation after solidification occurs on these nuclei. From the nitrogen analyses the critical concentration of aluminium nitride would appear to be that corresponding to about 0·016%N and this presumably represents the solubility of aluminium nitride in the liquid at some temperature just above the solidus.

The changes in the form of the precipitates in casts B2–B5 are not easy to explain as it would be expected that with increasing amounts of precipitate the grain boundary particles would become more continuous, giving relatively thick grain boundary films. Instead of this, the nitride changes its mode of growth to give needles extending into the austenite grains. It is possible that this process results from the segregation of aluminium atoms at the boundary with increasing aluminium content. From the metallographic observations it also appears that the grain boundary weakness leading to intergranular fracture can occur without there being any visible grain boundary precipitate, and it may be that this weakness is also the result of a segregation process.

The analytical results show that relatively small amounts of aluminium nitride can produce appreciable amounts of intergranular fracture in low alloy steel castings when given a suitable heat-treatment. For example, cast B2 has an ester halogen nitrogen content of 0·0051% and shows 80% intergranular fracture, and cast B1 has a trace of intergranular fracture with an ester halogen nitrogen content of only 0·0017%.

It has been shown that titanium and zirconium may be successfully used by themselves for deoxidation without giving rise to the intergranular fracture which is produced in the same steels by the addition of aluminium, but it appears that whereas titanium additions will prevent the development of intergranular fracture even in the presence of aluminium, this is not

the case with zirconium additions. This result is not entirely unexpected since the stability of the oxides and nitrides decreases in the order ZrO_2, Al_2O_3, TiO_2, ZrN, TiN, and AlN, and additions of titanium and aluminium would tend to give Al_2O_3 and TiN whereas zirconium and aluminium additions would tend to give ZrO_2 and AlN.

Since it has been shown in this and previous work that aluminium nitride precipitation occurs on slow cooling below about 1150°C it would be expected that all aluminium-killed steels would contain precipitates or segregates of aluminium nitride. The reason that intergranular fracture is not observed in wrought steels may be due to a combination of low internal stresses in the ingots, and hot working of the metal which would tend to break up the precipitates at temperatures where solution and re-precipitation of the aluminium nitride would not readily occur. It is interesting to note, however, that in an investigation into the causes of longitudinal panel cracking in ingots Desai[11] showed that high aluminium or high nitrogen contents encouraged this form of defect and that the ductility of En18 composition steels at temperatures between 500° and 800°C was progressively reduced as the aluminium nitride content was increased. This work demonstrates that in addition to producing a grain boundary weakness at room temperature, aluminium nitride, by reducing the high temperature ductility of the steel, may also give rise to intergranular failure at elevated temperatures. Desai also showed that the partial replacement of aluminium by titanium or vanadium improves the ductility of the steels at about 600°C but that partial replacement by zirconium was not effective in improving the ductility. This suggests that vanadium would also be effective in preventing intergranular fracture if used either in combination with or in place of aluminium.

There would appear to be little doubt that aluminium nitride is one of the principal causes of intergranular fracture in commercial steel castings but it is clearly not the only cause of this defect. A simple test to determine whether aluminium nitride is responsible for intergranular fracture in any particular case would be to determine whether it could be removed by heating to 1250°C and caused to reappear by slow cooling below 1150°C.

SUMMARY AND CONCLUSIONS

It has been confirmed that aluminium nitride can give rise to a form of intergranular fracture in steel castings, and that in aluminium-killed steels this defect can be minimized by keeping the nitrogen content of the steel as low as possible, by restricting the aluminium additions to the very minimum required for complete deoxidation, and by a high cooling rate after solidification. It has also been established that intergranular fracture can be avoided by deoxidizing with titanium or zirconium or by using titanium in combination with aluminium. Intergranular fracture due to aluminium nitride can be removed by heating at 1250°C but it reappears at the new austenite grain boundaries on slow cooling below about 1150°C. Aluminium nitride contents equivalent to about 0·002%N can give rise to intergranular fracture in low alloy steels if they are slowly cooled after solidification, but with aluminium nitride contents equivalent to about 0·016%N, nuclei of aluminium nitride are formed in the melt and subsequent precipitation in the solid steel occurs on these nuclei so preventing the development of the intergranular weakness which produces intergranular fracture. Although a grain boundary precipitate of aluminium nitride has been observed in severely embrittled steels, the occurrence of intergranular fracture is not always associated with a visible aluminium nitride precipitate.

ACKNOWLEDGMENTS

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Practical estimation of composition balance and ferrite content in stainless steels

L. Pryce and K. W. Andrews

SYNOPSIS

Estimation of small amounts of (delta) ferrite in stainless or heat-resisting steels is useful in connexion with control during rolling. Formulae are given for equivalent nickel and chromium contents to be used with a modified Schaeffler-type diagram indicating amounts of ferrite. The diagram is based on measurement of ferrite content in commercial casts. Sigma phase is also considered. This forms much more rapidly from ferrite than from austenite, some advantage being obtained by ensuring a fully austenitic steel at high temperatures. Complete avoidance imposes further restrictions which are indicated. 1842

INTRODUCTION

SOME GRADES of austenitic stainless or heat-resisting steel may contain small amounts of ferrite at rolling temperatures. In such cases difficulties in rolling may arise and it is desirable to be able to estimate the amount of ferrite and to control it. Similarly, in some grades of chromium–nickel steel for service at high temperatures the problem of sigma-phase formation may also arise. The authors and their colleagues have previously suggested the use of formulae which may be based on a phase diagram for the three major components (Fe, Cr, and Ni) and factors for other elements.[1, 2] In a recent review Ergang and others[3] deal at some length with these formulae. In view of this interest the present paper puts on record some improvements in the original formulae, especially the factors for manganese and molybdenum, and refers to developments in the application to ferrite control.

FERRITE FORMATION

It has been found convenient to refer to an equivalent nickel and chromium content and to use a modified Schaeffler-type diagram[4, 5] in which these quantities are plotted in rectangular coordinates and lines are drawn representing 0, 5, and 10% ferrite. For this diagram, which is shown in Fig.1, the following formulae are used:

18–8 *steel*

$$Cr' = Cr + 3Si$$
$$Ni' = Ni + \tfrac{1}{2}Mn + 21C + 11\tfrac{1}{2}N$$

18–8 Ti *steel*

$$Cr' = Cr + 3Si + 10Ti'$$
$$Ni' = Ni + \tfrac{1}{2}Mn + 21C'$$

Here Ti' = soluble titanium
C' = soluble carbon = about 0·03

The titanium appears to have 'fixed' practically all of the nitrogen but leaves some of the carbon. Hence:

$$Ti' = Ti - 4\,[(C - 0{\cdot}03) + N]$$

Experience has shown that the nitrogen in these steels varies somewhat according to the method of melting:

Scrap charges	0·025%N
Base mix charges	0·015%N

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Mr Pryce is with Samuel Fox and Co. Ltd and Dr Andrews is with the United Steel Companies Ltd at Swinden Laboratories.

18–8 Mo–Ti *steel*

$$Cr' = Cr + 3Si + Mo + 10Ti'$$
$$Ni' = Ni + \tfrac{1}{2}Mn + 21C'$$

(Ti' and C' as above)

18–8 Nb *steel*

$$Cr' = Cr + 3Si + 4Nb'$$
$$Ni' + Ni + \tfrac{1}{2}Mn + 21C'$$

C' = about 0·03 (as above) and

$$Nb' = Nb - 8\,[(C - 0{\cdot}03) + N]$$

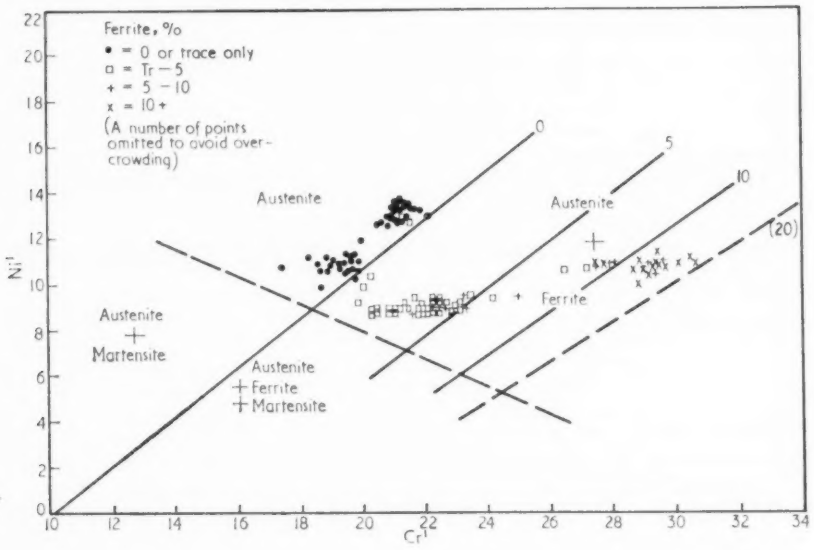
Molybdenum

The potency of molybdenum as a ferrite former appears to be dependent on three effects:

(i) it is not constant over a wide range and tends to vary with Mo content
(ii) it is affected by the presence or absence of other strong ferrite formers
(iii) it is probable that the factor for hot-rolled stainless or heat-resisting steels is not the same as, for example, the factor for weld-metal constitution.

These observations, particularly the last, may also apply to W, Nb, and Ti.

With Mo contents of 0·5–1·0% a factor of 1 appears to operate even with considerable amounts of Ti, but with higher Mo contents, up to 3%, the factor is larger and is probably as high as 4. It is therefore necessary to determine (e.g. by microscopical or magnetic methods) the amounts of ferrite for a number of casts within the range of composition for a specified molybdenum-containing stainless steel.



1 *Schaeffler-type diagram for amounts of ferrite at hot-working temperatures (~1 150°C)*

Manganese

It will be seen that the factor given for manganese is 0·5. In the previous work a factor of 1·5 had been proposed. Further work has shown this to be too high. The original choice had been made on the basis of the influence of Mn in low-alloy steels.[6] It is now appreciated that with more than about 12% chromium the potency of manganese as an austenite former is very much reduced. This effect is present in Fe–Cr–Mn alloys and is reflected in the shape of the austenite-phase boundary in that system.[7] The principle evidently extends to steels containing appreciable amounts of nickel and experience in our laboratories and elsewhere in connexion with high-manganese stainless steels reflects this tendency. Gow and Harder have suggested a factor of $\frac{1}{3}$, but this appears to be too low at least for the steels and conditions with which we have been concerned. These relationships are supported by extensive experience with works casts.

Figure 1 is based on measurements with some 170 casts for which the amount of ferrite was determined microscopically. Of these, 117 were definitely within the duplex austenite+ferrite region containing up to 20% ferrite. The following figures indicate the accuracy which can be expected in the duplex region:

Steel	No. of casts	Standard deviation
Cr 17–19, Ni 9–10, Ti (5×C)	94	1·68% ferrite
Cr 18–19, Ni 9–10, Mo 1·4–2 – Ti (5×C)	23	3·9% ferrite
Combined results	117	2·3% ferrite
Cr 17–18, Ni 10½–11½, Mo 2½	30	No ferrite or
Cr 16–18½, Ni 8–10	25	trace only

The deviations, based on differences between observed ferrite and amount from Fig.1 are comparable with those obtained elsewhere for weld metals.[5]

There are a variety of obvious possible causes which could account for the differences between observed amounts of ferrite and the amounts from Fig.1. The differences however, tend to be larger with larger amounts of ferrite and are not serious from a practical point of view.

SIGMA-PHASE FORMATION

The formulae given above indicate, in conjunction with the Schaeffler-type diagram, at what point a steel may contain a small amount of ferrite at high temperatures. Earlier work by the authors[1, 2] and others has shown that sigma phase forms much more rapidly from the ferrite phase than from the austenite matrix. Hence from this point of view a fully austenitic structure is desirable; at least it considerably extends the time required for the maximum effect of sigma formation to be reached.

However, it was indicated that for sigma-phase formation to be avoided altogether further limits must be set on the composition. On the basis of this work, but with the new factor for manganese, it appears that sigma formation in steels with 20–25%Cr and 15%Ni occurs under conditions represented approximately by the formulae:

at 900°C, if $Cr' > 17{\cdot}5 + 0{\cdot}37\ Ni'$
at 700°C if $Cr' > 17 + 0{\cdot}06\ Ni'$
(after long periods of heating)

In these formulae

$$Cr' = Cr + 3Si - 12\ (C - 0{\cdot}03)$$
$$Ni' = Ni + 0{\cdot}5Mn + 21 \times 0{\cdot}03$$

These expressions allow for C′=about 0·03%C left in solid solution and the rest combined as $(Cr, Fe)_{23}C_6$. For the avoidance of sigma-phase formation at temperatures below 700°C even lower chromium contents would be necessary.

APPLICATIONS

From a practical viewpoint the formulae for ferrite formation and Fig.1 can be used to deduce the amounts of ferrite from the chemical composition of the steels and thus to give a prior indication of their behaviour under hot rolling. Under these conditions the problem of sigma formation is not important. In steels with over 20%Cr or Cr equivalent, however, sigma could be a source of trouble. It becomes more

significant after service at high temperatures for long periods of time in Cr–Ni steels or in welds of Cr–Ni steels with 3%Mo.

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A radiochemical investigation of the combustion method for the determination of sulphur in cast iron

R. C. Rooney, A.R.I.C. and F. Scott, B.Sc., F.R.I.C.

SYNOPSIS

The combustion method for the determination of sulphur in cast iron has been examined, using S^{35} *as a tracer. The reasons for non-stoichiometry have been investigated, and the distribution of sulphur in the apparatus examined.* 1832

INTRODUCTION

THE COMBUSTION METHOD for the determination of sulphur in cast iron and steel is well known and widely applied, but suffers from the disadvantage of being non-stoichiometric. This calls for empirical standardization with samples that have been analysed by a referee procedure such as the British Standard gravimetric method,[1] and subsequent close adherence to the conditions of combustion etc., used for the standardization.

Many workers have published versions of the procedure,[2–9] and a wide range of conditions has been recommended. Suggested temperatures range from 1270°C, to 1440°C for unalloyed materials and 1300°C to 1460°C for alloyed steels and irons; some authors have used fluxes such as tin, copper, pure iron, vanadium pentoxide, or cobalt oxide, while others have used no flux. Oxygen flow rates have varied from 750 ml to 3000 ml/min; some workers filtered oxide dust from the gas stream with porous silica or firebrick plugs in the hot zone of the furnace, others used cotton or asbestos wool in the cold zone, and a further variation used a preheating period and no filters. Recommended absorbents include water, hydrogen peroxide, iodine solution, potassium iodate solution, or silver nitrate solution.

Smith and his co-workers[10] carried out an investigation of sulphur losses using an empty tube method with no flux and employing a preheating period, and drew a number of conclusions from the results obtained by altering some of the variables in the procedure. They were hampered by the difficulty of determining sulphur left in the boats and deposited on the furnace walls, so that some assumptions had to be made in the final calculations; in spite of these difficulties, useful results were obtained although the authors stress that the conclusions drawn may not necessarily apply when different experimental conditions are used.

In view of the variety of recommended conditions, it was decided that the conditions in routine use at the Alvechurch laboratories of The British Cast Iron Research Association should be used as a basis for investigation. These conditions have been in use for a number of years, and, in common with most of the recommended procedures, were arrived at empirically; it is not claimed that they have any outstanding advantage over other procedures, and in fact it was later found that they were inferior on some counts.

The procedure uses a flow rate of 1000 ml/min, a furnace temperature of 1330°C, no flux, and no preheating period. In view of the large amount of ferric oxide dust obtained when cast iron samples are burnt, a porous firebrick plug in the hot zone of the furnace is used to filter the gas stream, and the preferred absorbent is dilute iodine solution.

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TABLE I Typical sequence of figures obtained with tube stabilized by burning 10 samples

Temp., °C	Flow rate, ml/min	Yield $\%SO_2$	Mean
1200	1000	62·8; 60·0; 60·4; 60·0	61
1200	2000	66·9; 66·6; 66·9; 67·0	67
1200	3000	64·2; 64·8; 64·0; 64·2	64
1300	1000	69·7; 70·0; 70·3; 69·8	70
1300	2000	73·5; 75·2; 74·8; 75·6	75
1300	3000	70·7; 70·4; 70·0; 69·8	70
1400	1000	75·2; 75·8; 76·6; 76·0	76
1400	2000	80·7; 79·3; 80·2; 79·4	80
1400	3000	75·2; 74·0; 76·0; 75·3	75

In order to examine the reasons for non-stoichiometry it would appear to be desirable to investigate the distribution of sulphur in the combustion apparatus; the use of radioactive sulphur and tracer techniques was an obvious choice. An investigation of a similar nature has been reported by Fulton and Fryxell[11] since the inception of this work, but they used radio-frequency induction heating in a vertical tube furnace; these conditions are not widely employed in the UK. In addition, they used as a radioactive sample a mixture of iron powder and labelled ferrous sulphide, and it was considered possible that this mixture would not necessarily give the same results as a cast iron sample containing sulphur-35 as a constituent. Finally, the actual amount of tracer work reported was small, relating apparently to a single experiment, and not presented in detail. It was decided to prepare a properly labelled sample and to use it in a systematic investigation.

EXPERIMENTAL

Experiments with inactive sulphur

As a preliminary to the radio-chemical work, an investigation of variation of yield of sulphur with temperature and flow rate was made, using the apparatus of the BCIRA methods of analysis subcommittee.[7] Samples (1 g) of BCS standard cast iron no.234/4 (sulphur content 0·113%) were burnt, and the yield of SO_2 determined iodimetrically. The standard sodium thiosulphate solution was standardized against potassium iodate using the procedure given by Vogel,[12] and the yield of SO_2 calculated on the basis of the reaction

$$H_2SO_3 + I_2 + H_2O \rightarrow H_2SO_4 + 2HI$$

Temperatures of 1200°C, 1300°C, and 1400°C were used, with oxygen flow rates of 1000, 2000, and 3000 ml/min. Attempts to use temperatures higher than 1400°C were prevented by consistent failure of the Alpolain combustion thimbles.

The furnace temperature was checked with a Pt/Pt–Rh thermocouple before, and after, each sample and the oxygen flow rate adjusted by means of a Rotameter flow meter immediately after insertion of the boat. Some fluctuation of flow rate occurred while the sample was burning, but no correction was made as the fluctuations were small and rapid.

Various bubblers and absorbents were used, but no significant advantage was found with any particular combination. The bubblers used were sintered glass dispersal tubes, perforated bubblers of the Westwood and Mayer type and simple glass jets. The absorbents were iodine solution, potassium iodate solution, N sodium hydroxide, and water. It appears that the solubility of SO_2 in water is sufficiently high to override considerations of comparative bubbler efficiencies etc., although these would probably become important if alkalimetric titration was used, when some sulphur trioxide is also titrated. Experiments using second traps failed to reveal any sulphur dioxide escaping absorption, even with the very low yield of sulphur dioxide obtained at 1400°C and 3000 ml/min.

TABLE II Composition of S^{35} labelled cast iron sample

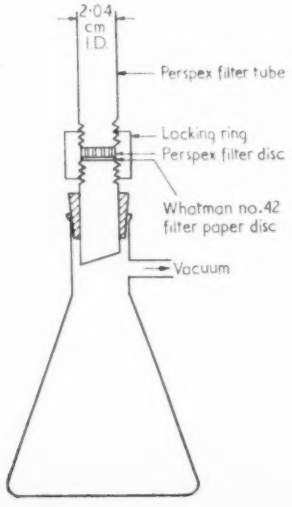
%C	%Si	%S	%P	%Mn
3·46	2·43	0·076	0·030	0·58

Serious variations were caused by altering the combustion apparatus. It is well known that the yield drops as the furnace tube ages, and Smith and his coworkers ascribed this to catalysis of the oxidation of sulphur dioxide to trioxide by the ferric oxide coat on the tube wall. It was found that a new tube would often give a different yield from the previous tube when new, and two firebrick plugs often gave different yields. Most sets of apparatus reached a fairly steady state between 70% and 80% yield at 1330°C and 1000 ml/min after about 10 samples had been burnt, and thereafter the yield changed very slowly as build up of ferric oxide took place. A typical sequence of figures obtained with a tube stabilized by burning 10 samples is shown in Table I.

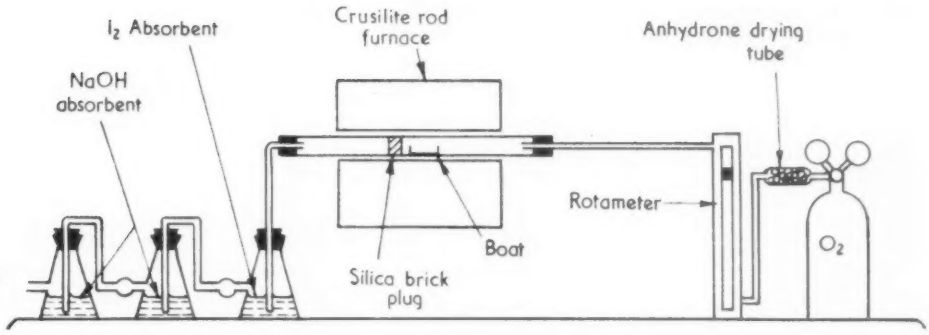
Preparation of radioactive sample

Sulphur-35 is available from the Radiochemical Centre, Amersham, as labelled ferrous sulphide, but this form is relatively expensive. A cheaper source of supply is sulphur-32 irradiated at Harwell, although the product is of lower specific activity and may be contaminated with phosphorus-32. A sample of 20 g of sulphur was submitted to AERE Harwell, and irradiated for 14 days at a pile factor of 14 in BEPO.

The sample was examined on receipt and found to be badly contaminated with phosphorus-32; this was probably due to the high pile factor, as the reaction $S^{32}(n, p)\ P^{32}$ is a fast neutron reaction. It was necessary to separate the phosphorus as the respective



1 *Filtration apparatus*



2 *Combustion and absorption apparatus*

specific activities were S^{35} 1·14 mc/g, and P^{32} 4·4 mc/g; the sample residues in the boats were to be examined, so that the phosphorus would cause serious interference.

Attempts to remove the phosphorus by oxidation with nitric acid or aqua regia were unsuccessful, and the product was purified by converting the sulphur to ferrous sulphide, liberating hydrogen sulphide and using the gas to precipitate ferrous sulphide from ammoniacal ferrous tartrate solution. The precipitate was filtered, washed, and dried, and found to be radiochemically pure when an absorption curve was plotted.

5 g of this labelled ferrous sulphide was packed into a mild steel capsule with a screw-in cap, and the capsule added to 4 lb of Swedish pig iron, with ferrosilicon and ferromanganese additions to produce a typical grey iron composition. The melting was carried out at the BCIRA Alvechurch laboratories, in the National Research Corporation hf vacuum melting and casting furnace, under a pressure of 0·5 atm of argon, and a small ingot cast; the pouring temperature was 1450°C.

The ingot was analysed and found to have the composition shown in Table II. The sample was taken by drilling with a ½ in. drill at 150 rev/min with due precautions against radiation hazards.

The sulphur was determined on samples taken from a number of points on the ingot, and the ingot shown to be reasonably homogeneous. The figures obtained were: 0·076,* 0·077,* 0·075,* 0·078,* 0·075,† 0·077,† 0·076,† 0·075%S.†

Radiochemical techniques

S^{35} is a weak β emitter (energy 0·167 MeV) so that the personal radiation hazard in this work was low. It was found necessary to use solid counting techniques after precipitation of barium sulphate, in order to achieve the desired sensitivity, and a method similar to that of Collins and Myers[13] was used.

In counting, 0·02 g/cm² represents infinite thickness so that by maintaining the pad thickness at this level, maximum sensitivity and precision are obtained. The Geiger Müller tube window was 2·04 cm dia., so that a pad of 3·3 cm² was required, i.e. 66 mg. This represents 9·05 mg of sulphur as carrier, so that the error introduced by the additional sulphur derived from the sample (always less than 0·75 mg) was small.

* Gravimetric figures.
† Combustion figures.

A calibration curve for the counting equipment was constructed by dissolving some of the labelled sample in aqua regia, and taking aliquots equivalent to 0, 20, 40, 60, 80, and 100% of the 1 g sample to be burnt. Then 25·0 ml of a solution of K_2SO_4 (1·965 gpl), which corresponded to the desired amount of carrier, was added and the solutions processed as for the standard gravimetric procedure.[1] The sulphur was precipitated by the addition of 25 ml of barium chloride solution (3·0 gpl) to the boiling sample solution, the precipitate allowed to settle for 4 h, and filtered on the apparatus shown in Fig.1. This controlled the weight and size of the barium sulphate pad satisfactorily. After washing with water and alcohol, the filter paper disc was mounted on a planchette, dried for 30 min and counted. The counting equipment consisted of a Mullard G.M.4 mica window GM tube, of window thickness 1·7–2·3 mg/cm², a Panax probe unit and IDL EHT unit, scaler, and prescaler. The sample was mounted as close to the tube window as possible.

The counts per minute (cpm) were corrected for background (7–9 cpm) and lost counts, and a calibration curve plotted. This was a straight line through the origin, with a slope such that 1000 cpm was equivalent to 6% of the total sulphur. This enabled as little as 0·03% of the total sulphur to be determined. To allow for decay, the 1 g standard was thereafter counted daily, but the variation was so small as to be negligible.

The fact that the calibration curve is a straight line may be taken as an indication that any inaccuracies due to there not being any marked excess of barium ions present when the precipitation was carried out are inside the limits of experimental error. It was considered advisable to use a minimum of barium chloride, in order to avoid the possibility of adsorbed barium in the pad, which would lead to very strong self-absorption of the weak β-rays.

Samples (1 g) of the S^{35} labelled iron were burnt at 1300°C and 1400°C, using flow rates of 1000 and 2000 ml/min. The iodine absorbent contained 25 ml of K_2SO_4 carrier, and was titrated with thiosulphate in the usual way to obtain the chemical yield. The sulphur introduced in this titration was negligible. After combustion, the various fractions were treated as follows:

Absorbent Rendered just acid with hydrochloric acid, boiled, and barium chloride added. (The delivery tube was rinsed into the absorbent with water.)

Boat Cooled, and treated in a platinum basin with 25 ml of carrier solution and a mixture of hydrofluoric, hydrochloric, and perchloric acids. This

TABLE III Distribution of sulphur in combustion and absorption train

Fraction	Total sulphur, % 1300°C 1000 ml/min	1300°C 2000 ml/min	1400°C 1000 ml/min	1400°C 2000 ml/min	1400°C 3000 ml/min	Approx. total counts	Approx. counting time, min	Approx. accuracy, %
$\%SO_3$ by titration	80	82	84	87	68	...	...	...
Iodine	82·7	84·8	88·9	90·1	68·7	2×10^5	15	0·2
NaOH (1)	3·9	5·0	3·7	3·9	27·6	1×10^4	15	1
NaOH (2)	0·61	2·2	0·27	0·06	0·22	5×10^3	45	1·5
Boat	0·71	0·87	0·14	0·14	<0·03	5×10^3	45	1·5
Plug	3·3	0·56	0·58	0·47	0·63	5×10^3	20	1·5
Tube	8·3	5·4	5·9	5·2	2·9	1×10^4	10	1
Total	99·52	98·83	99·49	99·87	100·05	...	...	...

mixture dissolved out practically all the slag residue. The solution was evaporated to fumes of perchloric acid, diluted, filtered, and barium sulphate precipitated.

Silica brick plug This was treated in the same way as the boat.

Tube This was removed from the furnace, rinsed into a 500 ml conical beaker with concentrated hydrochloric acid and 25 ml of carrier solution added. The tube was then supported in the beaker and the acid boiled so that it refluxed inside the tube for 15–20 min. The tube was removed, rinsed with water and the solution evaporated to low bulk, diluted, filtered, and barium sulphate precipitated.

The first results obtained using this technique indicated a loss of active sulphur, so a further series of tests was carried out, in which the gases from the absorbent were led into a second absorbent vessel and bubbled through 20% sodium hydroxide plus carrier. The white fumes which were seen to pass through the first absorbent appeared to be completely retained by the second. This second absorbent was acidified with hydrochloric acid, carbon dioxide boiled out, and barium sulphate precipitated. The results were still slightly low, so a third series of tests was carried out using a third absorbent vessel, again containing 20% sodium hydroxide, plus carrier. The final apparatus is shown in Fig.2. The results for iodine absorbent, boat, plug, and tube showed good agreement in these three sets of tests.

The decrease in yield of SO_2 noted in the inactive experiments when a flow rate of 3000 ml/min was used, was investigated by burning a sample at 1400°C and this flow rate.

RESULTS

The results for the final series of experiments are given in Table III.

The accuracy with which the various fractions can be determined is dependent upon the complete recovery of sulphur from the apparatus and on the counting accuracy. The counting accuracy is inversely proportional to the square root of the total number of counts; for 10000 counts the accuracy is $1/\sqrt{10000}$, i.e. 1/100 or 1%. By counting for a long enough period, even low count rates can give quite high accuracies. An approximation of the accuracy of the counting is given in Table III.

DISCUSSION

It will be noted that the chemical yield figures for sulphur dioxide are higher in Table III than in Table I. This discrepancy is larger than would be expected from differences in tube and plug, and is probably due to the use of a different furnace for the two series of experiments. The platinum-wound furnace used for the inactive experiments failed during the work, and was replaced by a 'Crusilite' furnace with a longer heated zone. The significance of this is discussed later.

The results presented in Table III show that provided the temperature is sufficiently high, less than 1% of the sulphur is retained by the residue in the boat and the filter plug. 1300°C appears to be near the lower temperature limit, as at this temperature the plug at 1000 ml/min retained 3·3% of the sulphur.

The difference between the chemical yields of sulphur dioxide and the radiochemical figures for the iodine absorbent are presumably due to the retention of some sulphur trioxide. The differences at 1000 and 2000 ml/min agree with the figures for retention of trioxide reported by Smith *et al.*

The figures show that the major sources of loss are:

(i) sulphur trioxide condensing or being adsorbed on to the walls of the apparatus

(ii) sulphur trioxide escaping absorption by the dilute aqueous absorbents normally used.

The furnace tubes were the most difficult parts of the apparatus to clean, and it is considered probable that the sulphur not accounted for was not washed from the tube wall.

In order to minimize the loss of sulphur, it would be necessary to minimize the formation of sulphur trioxide, as it has been shown that the dioxide can be quantitatively absorbed at flow rates up to 3000 ml/min.

The low results obtained at 3000 ml/min are not in accordance with the figures of Smith *et al.*, and this is probably a pressure effect according to Le Chatelier's principle. Smith used an open tube, which would require little or no backing pressure to obtain high flow rates. With a sealed plug, it is necessary to use a pressure of up to 15 lb/in² to obtain a flow rate of 3000 ml/min. The pressures required for flow rates of 1000 and 2000 ml/min were 2 lb/in² and 6 lb/in² respectively.

The recovery of sulphur dioxide is very low compared with the composition of the gases at equilibrium calculated from thermodynamic data (3%SO_3 at 1300°C and 2%SO_3 at 1400°C).

Smith showed that losses of sulphur dioxide of about this level occurred if sulphur dioxide/oxygen mixtures were passed through clean tubes at 1300°C, which indicates that equilibrium is attained rapidly. If it is assumed that equilibrium is attained at the sample, which is in the hottest part of the tube, then the major

source of sulphur trioxide must be reaction of dioxide and oxygen in the cooler parts of the tube. At temperatures below 300–400°C the reaction, even if catalyzed, becomes very slow, so that the reaction zone can be defined as that part of the tube between the sample and the point where the gas temperature has fallen to about 300°C. The increase in yield of sulphur dioxide with temperature can then be regarded as being due partly to the steeper temperature gradient along this zone, so that less time is available for reaction, and partly to less trioxide being present at equilibrium; the increase in yield of dioxide with increased flow rates up to 2 l./min is then due to higher linear gas velocities through the reaction zone, again reducing the time available for reaction. The use of a furnace with a longer hot zone increased the yield of dioxide, which supports this theory.

Smith ascribes the decrease in yield of dioxide as the tube ages to catalysis of the dioxide/trioxide oxidation by ferric oxide dust in the cooler parts of the tube; this would be of less importance if the gases were in contact with this part of the tube for as short a time as possible, and if the temperature were raised. This might be achieved if the filter plug were omitted, so exposing the ends of the tube to direct radiation from the hot zone, but this would call for some method of suppressing the large amount of ferric oxide dust which is formed when cast iron samples are burnt. Omission of the filter plug would also permit the use of higher flow rates without the use of high backing pressures.

CONCLUSIONS

It has been shown that the major losses of sulphur in the combustion method are due to condensation of sulphur trioxide in the apparatus, and to loss of sulphur trioxide which passes through the aqueous absorbent. Very little sulphur is retained in the residue, or on the filter plug, provided that the temperature is sufficiently high.

ACKNOWLEDGMENTS

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STAL *in English*

No 4 of the 1960 series of *Stal in English* has been published and subsequent issues will appear at about three-weekly intervals. Full details of subscriptions to *Stal in English* are available from the Secretary of The Iron and Steel Institute.

The notch ductility and tensile properties of some synthetic mild steels

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SYNOPSIS

A detailed study has been made of the effect of nitrogen on the Charpy V-notch transition curve, and on the tensile properties at +20°C and −192°C of a 0·17%C–0·8%Mn steel made from high-purity materials when air-cooled from 880°C. The effect of aluminium, silicon, zirconium, and titanium on a similar steel containing 000·7%N has also been studied. In addition, the effect of air cooling from other temperatures in the range 880–1200°C on the Charpy V-notch properties has been examined.

Nitrogen up to 0·010% progressively broadens the temperature range over which the fracture of Charpy specimens transforms from fully brittle to fully tough (transition range). The highest temperature at which a fully crystalline fracture occurs (transition temperature) is unaffected.

Aluminium, zirconium, and titanium, which form nitrides, narrow the transition range to a few degrees. Aluminium up to 0·08% lowers the transition temperature, owing to (a) *its effect on the ferrite grain size and* (b) *its solubility in the ferrite. A rise in transition temperature is produced by raising the aluminium from 0·08 to 0·1% and is associated with an irregular grain size. Further additions, however, lower the transition temperature by increasing the aluminium in solid solution.*

Titanium refines the ferrite grain size and lowers the transition temperature. Zirconium has no effect on grain size and little effect on the transition temperature. Silicon has no significant effect on the transition range but up to 0·3% lowers the transition temperature. Further additions of silicon, up to 1·0%, progressively raise the transition temperature.

A correlation has been established between the transition temperature and the fracture stress at −192°C for most of the steels, although some anomalies have been found. Variations in the austenitizing temperature within the range 880–1200°C have no effect on the transition curves for the steels containing nitrogen, and no marked effect on the grain size. In steels containing aluminium, raising the austenitizing temperature raises the transition temperature and increases the ferrite grain size. In steels containing titanium, raising the austenitizing temperature has no effect on the transition curve and no significant effect on the grain size. In steels containing zirconium the transition temperature is raised slightly and is accompanied by a small increase in grain size.

Details of the preparation of the 'synthetic steels' from high purity materials, and the calculated distribution of nitrogen and aluminium are given in the appendixes. 1792

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INTRODUCTION

THE BRITTLE FRACTURE of mild steels at low temperatures was reported as early as 1879[1]; many instances have been recorded since this date and those occurring up to 1953 were surveyed by Shank.[2]

Since 1936 several serious catastrophic failures of mild steel structures including bridges, ships, and oil storage tanks, have stimulated considerable research into the causes of brittle fracture in structural steels.

Previous work has drawn attention to the importance of design and welding in determining the susceptibility of steels to this type of failure, and other work such as the examination of US ship plate steels carried out by Williams[3] has indicated that the notch ductility of commercial steel can vary over wide limits. Some of these variations can be attributed to differences in composition; the effects of carbon and manganese have been fairly well assessed and correlations have been established between some other elements and notch ductility. The results of different workers, however, are conflicting and little attention has been paid to the possible effect of interaction between some of these elements.

It has also been shown that variations in rolling conditions, in particular the temperature of the steel during the last few passes, can affect the notch ductility and normalizing has been shown to improve many steels. Some of the observed variations in notch ductility between commercial steel plates of the same nominal composition and even between parts of the same plate, however, cannot be explained adequately on the basis of previously published work.

A consideration of some of the differences which occur in steelmaking practice suggests that some of the variations in notch ductility of structural steels may be associated with the minor elements present. These elements can be divided into two groups (*a*) those introduced or affected by deoxidation practices, and (*b*) those arising from the raw materials used in the manufacture of mild steel. The work described in this report is confined to an investigation of the effect of some of the elements in the first group. In order to avoid complications which could arise through variations in the amounts of elements of the second group, however, the research was carried out on Fe–Mn–C alloy made from high-purity materials to which the element or elements concerned were added. The base 'synthetic steel' was designed to be of a similar nominal composition (i.e. with respect to manganese and carbon) to a commercial steel.

The additions made to the synthetic steels consisted of nitrogen alone in various amounts, and a constant amount of nitrogen (0·007%) together with varying amounts of aluminium, silicon, zirconium, or titanium. The oxygen content of the synthetic steels was kept to a constant low level of less than 0·001% by vacuum melting and hydrogen deoxidation, so that such effects which might result from the formation of oxides of the added elements would be kept to a minimum.

In the steels containing nitrogen only, the amounts varied from 0·002% to 0·016% to cover the range found in commercial converter, OH, and electric steels. Aluminium or silicon, the most common deoxidants employed in steelmaking, were added with nitrogen to other steels. Zirconium and titanium, which, like aluminium are strong nitride formers, were added with nitrogen to a further series of steels.

The importance of rolling conditions and the subsequent thermal history of the steel has been indicated in previous work and the rolling schedule for the synthetic steels was, therefore, standardized. After rolling, all the steels were austenitized and air cooled from various temperatures within the range 880–1200°C before testing.

Numerous tests have been suggested for assessing the susceptibility of a steel to brittle fracture. The Charpy test using a specimen with a V-notch has been widely adopted in the UK for purposes of control and the Charpy V-notch values are included in some British Standards for structural steels. The test is simple to carry out, employs an easily machined test piece, and is economical with respect to the material. Primarily it measures the resistance of a steel to the initiation of a brittle crack and no other test suggested for this purpose is as convenient. The Charpy test was, therefore, adopted for assessing the resistance of the synthetic steels to the initiation of a brittle crack. Differing views are held on the relative importance of employing a measurement of the energy absorbed during fracture and of the ratio of the crystalline and fibrous areas in the fractured surface as criteria for measuring the notch ductility of a steel. Both have been adopted by different investigators in the past. In this investigation both criteria are reported so that the results can be compared with other work. A third criterion, which is considered to be less affected by the resistance of the steel to the propagation of the crack, was also used.

In addition to studying the effect of the added elements on the notch ductility of mild steel, their effect on the tensile properties of the steels air cooled from 880°C has also been examined. Tensile tests at +20°C and −192°C have been carried out and attempts have been made to correlate the tensile and Charpy properties.

PREVIOUS WORK

Effect of steel composition on notch ductility

Nitrogen

Dickie[5] showed that the Izod energy transition for air-blown basic Bessemer steel occurred at a lower temperature when modifications were made to the steelmaking process which reduced the nitrogen content. Enzian and Salvaggio[6] also showed that higher nitrogen contents of a commercial semi-killed BOH steel gave higher Charpy V-notch transition temperatures.

Rinebolt and Harris,[7] however, working with laboratory alloys deoxidized with ferrosilicon and aluminium showed that increasing the nitrogen content up to 0·015% progressively lowered the Charpy keyhole 10 ft lb transition temperature, but that further increases raised it.

Aluminium

Rinebolt and Harris[7] have also shown that increasing the soluble aluminium content of laboratory produced steels, from 0·001% to 0·019%, lowers the 15 ft lb Charpy keyhole transition temperature but that further increases have no effect. On the other hand, Geil *et al.*[8] showed that increasing the aluminium content of a 1·0%Mn–0·09%Cr–0·3%C steel in the quenched and tempered condition lowered the transition temperature to a minimum, and that further increases raised it.

Crafts and Ollenhauer[9] found that increasing the soluble aluminium content of a plain carbon steel (of unspecified carbon content) up to 0·2% lowered the 15 ft lb Charpy keyhole transition temperature and that further additions up to 0·4% had no effect. Frazier *et al.*[10] made a detailed study of the effect of aluminium additions (up to 0·08%) on the 12 ft lb

TABLE I Chemical composition of 'synthetic steels' containing additions of aluminium and nitrogen

Steel no.	PA	PM	PR	PC	PG	PS	PD	PE	PN	PO	PP	PQ
C, %	0·15	0·17	0·17	0·16	0·17	0·16	0·17	0·19	0·21	0·019	0·18	0·19
Mn, %	0·68	0·78	0·77	0·86	0·72	0·69	0·68	0·80	0·81	0·77	0·85	0·78
Al, % (insoluble)	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001
Al, % (soluble)	0·0015	0·0018	...	0·001	0·001	0·0014	0·03	0·08	0·111	0·193	0·456	0·995
N, %	0·002	0·0032	0·0047	0·007	0·010	0·016	0·007	0·0078	0·0078	0·0073	0·0062	0·0062
Oxygen, %	0·0010	0·001	0·0008	0·0005	0·0005	0·0009	0·0006	0·008	0·0009	0·0012	0·0015	0·0015
S, %	0·01	0·01	0·01	0·01	0·01	0·01	0·01	0·01	0·01	0·01	0·01	0·01
P, %	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001	0·001
Si, %	0·01	0·026	0·013	0·01	0·01	0·016	0·01	0·01	0·016	0·014	0·014	0·011

Charpy keyhole and Navy tear test transition temperatures of a series of laboratory casts of a 0·2%C semi-killed steel having two levels of manganese and three levels of silicon. They showed that the addition of aluminium lowered the transition temperatures but the effect was less in the steels containing 0·5%Mn than in those containing 0·8%. They also showed that the effect was less in steels having the higher levels of silicon.

Silicon

The results given in published work on the effect of this element appear to be in fairly good agreement. Rinebolt and Harris[7] showed that up to 0·3% silicon lowered the 15 ft lb Charpy keyhole transition temperature of a 0·3%C aluminium-killed steel but that further additions raised it. Frazier *et al.*[10] obtained similar results from laboratory steels containing various levels of manganese and aluminium, but also found that the silicon addition corresponding to the minimum in the transition temperature varied with the manganese content. They attributed this minimum to deoxidation of the steel by the silicon.

Some of the discrepancies between the results obtained by different workers described above could be due to interaction between the elements in the steel or the fact that the steels have been examined in different heat-treatment conditions. Full details of composition and heat-treatment have, however, not always been given.

Effect of microstructure and rolling conditions on notch ductility

The effects of composition on the notch ductility of mild steel cannot be considered in isolation from the effects of microstructure. It has been shown that the notch ductility of steels can be influenced by variations in the conditions under which a steel is rolled and by heat-treatment. Hence variations in composition may affect the notch ductility of steels indirectly by changing the response of a steel to rolling conditions and heat-treatment. Miller[11] showed that, while steels deoxidized with silicon showed a progressive increase in austenitic grain size with increasing austenitizing temperature, steels deoxidized with aluminium showed no grain growth up to a certain temperature, but above this temperature rapid grain growth occurred. Other investigators[12, 13] have shown that this temperature increases with (acid-soluble) aluminium to a maximum at about 0·04–0·08%Al. The grain refinement produced by small amounts of aluminium has been attributed to aluminium nitride acting as nuclei or as a barrier to grain growth, but the maximum in the grain coarsening temperature/aluminium relationship has not been fully explained. The importance of grain size on notch ductility has been observed by many workers. Vanderbeck[14] for example, showed that there was a linear relationship between the ASTM grain size (obtained by heat-treating a mild steel at various temperatures) and the Charpy transition temperature. Mackenzie[4] has shown a correlation between finishing rolling temperature and ferrite grain size which corresponds with variations in Charpy V-notch transition temperature and Van der Veen[15] has stated that commercial steels of improved notch ductility are obtained by rolling at low temperatures. Mackenzie considers that the fine ferrite grain size is produced by a fine austenite grain size, which is controlled by the rate of grain growth at the rolling temperature. He also observed that increases in finishing rolling temperatures above 1100°C have little effect on notch ductility.

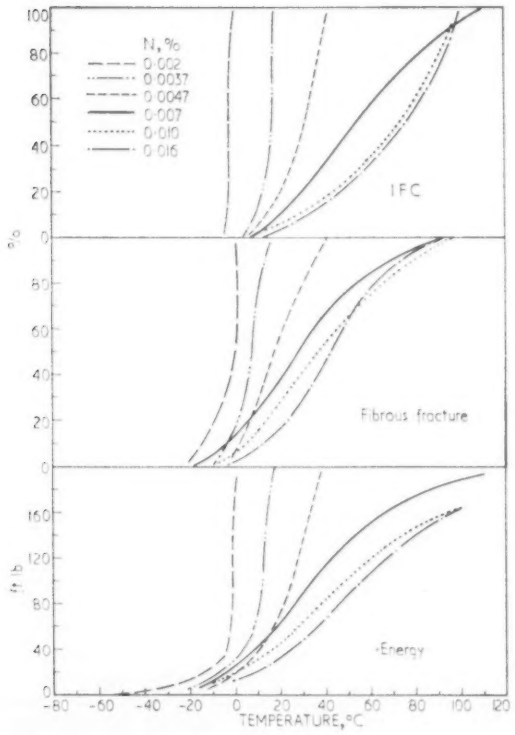
Frazier *et al.*[16] rolled a series of commercial plates under laboratory conditions and found that increasing the finishing rolling temperature increased the ferrite grain size and raised the transition temperature irrespective of the aluminium content; they reported more rapid grain growth in some steels than in others. Their results cannot, however, be explained adequately as they did not record the nitrogen and oxygen contents of the steels.

Burns and Judge[17] who studied the effect of subsequent heat-treatment on the structure and Charpy V-notch transition temperature of a commercial steel, concluded that while the transition temperature was partly related to ferrite grain size, it was also a function of the austenite grain size. They considered that pearlite in the structure had a secondary influence, slow cooling rates through the transformation range causing coarse pearlite structures and a higher transition temperature.

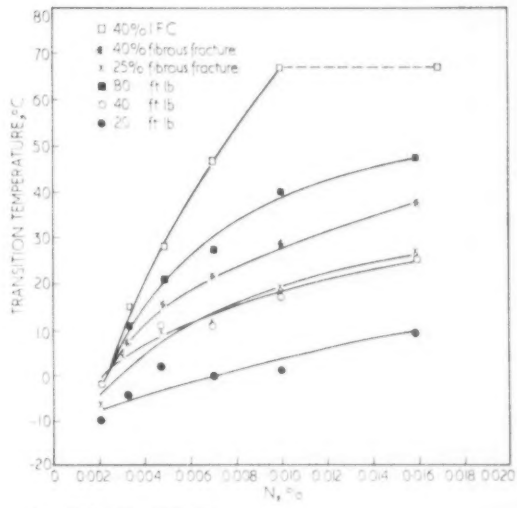
EXPERIMENTAL

Preparation of 'synthetic steels'

The 'synthetic steels' were made from a special quality Swedish iron, electrolytic manganese and spectrographic electrode carbon, by a procedure similar to that adopted by Hopkins *et al.*[18] for the manufacture of iron of high purity. The oxidation treatment given to the Swedish iron by Hopkins *et al.* was, however, omitted as removal of carbon and manganese from the iron was not necessary. To avoid possible contamination of the steels by aluminium a rammed magnesia crucible was substituted for the alumina crucible.



1 *Charpy V-notch transition curves for steels containing nitrogen, air cooled from 880°C*



(*a*) 20°C (*b*) −192°C

2 *Effect of nitrogen on the transition temperatures of steels air cooled from 880°C*

The iron was deoxidized with hydrogen and the nitrogen was introduced by holding the iron at constant temperature in contact with a nitrogen atmosphere at a controlled pressure. The added elements including carbon were introduced just before the steel was cast into 50-lb 4-in. square ingots in an argon atmosphere. The ingots were forged to 2-in. square and thereafter rolled to $\frac{5}{8}$-in. dia., under carefully controlled conditions.

The compositions of the steels are given in Tables I–III and full details of their preparation are given in Appendix I.

Heat-treatment

All the steels were tested after air cooling from 880°C and most steels were tested in at least three other conditions involving air cooling from temperatures in the range 880–1200°C. In all cases $\frac{5}{8}$-in. dia. bars of the steels were soaked at the austenitizing temperature for $\frac{1}{2}$ h. About 8 weeks elapsed between the heat-treatment and the testing of the steels except in the experiments on the effect of quench ageing described below.

The possible influence of quench ageing on the notch ductility of steels in the form of $\frac{5}{8}$-in. dia. bars was examined for two of the steels. For this purpose, some of the bars were air cooled from 880°C and tested within 24 h and others were water quenched from 700°C and tested within 24 h.

Charpy tests

The notch ductility of the steels was assessed from the transition curves plotted from the results of Charpy V-notch tests carried out at various temperatures. The specimens were 55 mm long × 10 mm square having a V-notch, 2 mm deep, an angle of 45° between the sides, and a radius at the root of 0·25 mm. Each notch was machined in one cut without lubrication with a specially radiused milling cutter. All the tests were carried out on a 30 kg Losenhausen impact testing machine.

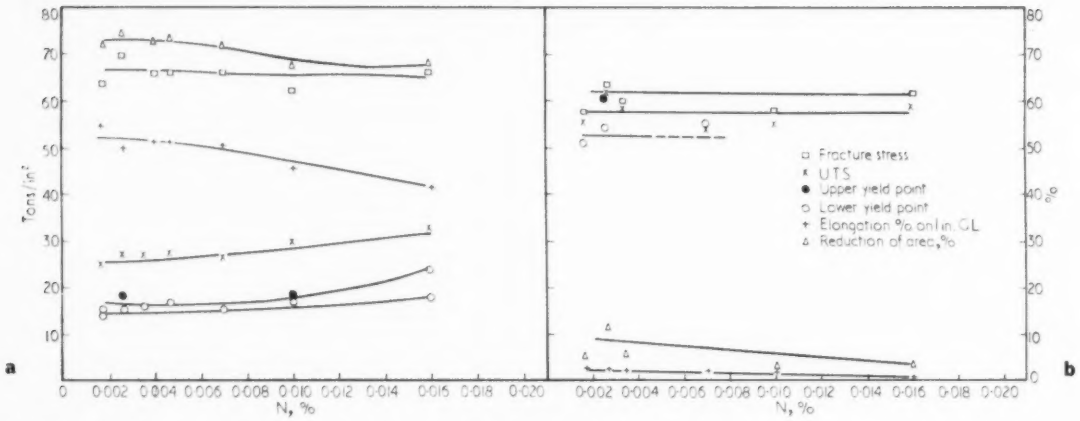
Specimens tested at temperatures between 0°C and room temperature were cooled in an ice–water mixture and those tested at lower temperatures were cooled in a mixture of solid carbon dioxide and acetone or alcohol. The specimens were placed in the coolant for a minimum of 15 min, transferred to the testing machine with cooled tongs and broken within 7 s after removal from the cooling mixture. Specimens treated

TABLE II Chemical composition of 'synthetic steels' containing additions of silicon and nitrogen

Steel no.	PF	PK	PU	PV
C, %	0·17	0·17	0·17	0·16
Mn, %	0·89	0·82	0·180	0·87
Si, %	0·03	0·30	0·45	0·93
N, %	0·008	0·008	0·0085	0·007
Oxygen, %	0·0007	0·0011	0·0012	0·0014
Al, % (soluble)	0·005	0·0015	...	...

TABLE III Chemical composition of 'synthetic steels' containing additions of titanium and zirconium

Steel no.	PW	PX	PY	PZ
C, %	0·16	0·17	0·17	0·16
Mn, %	0·80	0·76	0·75	0·81
Ti, %	...	...	...	0·02
Zr, %	0·016	0·048	0·057	...
Ni, %	0·0078	0·008	0·0078	0·008
Oxygen, %	0·0017	0·0014	0·0008	0·0009
Si, %	0·010	0·008	0·016	0·025



3 *Effect of nitrogen on tensile properties of steels air cooled from 880°C*

above room temperature were heated in water or in an oil bath stabilized at the testing temperature. To minimize any possible ageing the specimens were kept in the oil or water for 10 min only.

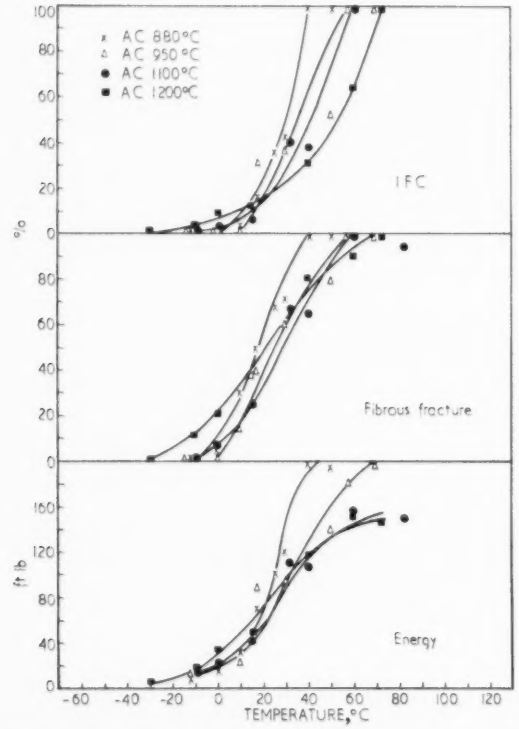
In most cases eight specimens were used for the determination of a transition curve. Three curves, based on the following criteria, were drawn for each steel in each heat-treatment condition:

Energy The first curve was based on the energy absorbed during the fracture of the Charpy specimen.

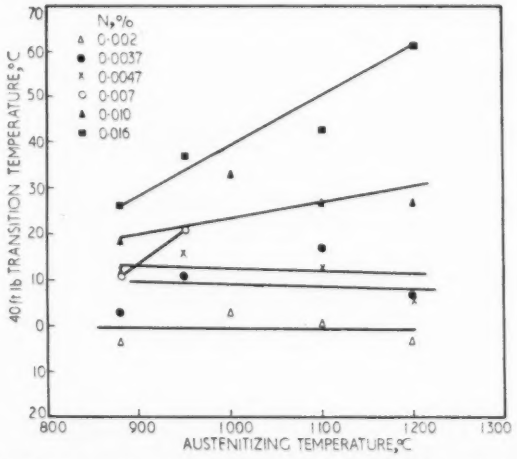
Fibrous fracture The second curve was based on the percentage of fibrous fracture in the fractured surface of the specimen. This was expressed as a percentage of the cross-sectional area of the specimen after fracture. Photographs of the fractured surfaces showing a mixture of fibrous and crystalline surfaces were obtained at a magnification of ×10 and the percentage fibrous fracture determined from planimeter measurements of the total area and the area of crystalline fracture.

Initial fibrous crack The energy and percent fibrous fracture (or percent cleavage fracture) criteria have been employed by previous workers but these values are influenced by the resistance of the steel to the propagation of the fracture as well as its resistance to the initiation of a crack. On the assumption that the fibrous crack progresses from the base of the notch across the specimen and the length of this crack is unaffected by the resistance of the steel to the propagation of a brittle crack, the length of the initial fibrous crack was adopted as a third criterion. This was measured from the photographs of the fractured surfaces and expressed as a percentage of the original depth of the specimen below the notch (8 mm).

In the discussion of the results in the following



4 *Charpy V-notch transition curves for steel containing 0·0047%N*



5 *Effect of austenitizing temperature on the 40 ft lb transition temperature for steels containing nitrogen*

TABLE IV Austenitic grain sizes of synthetic steels, grains/in² × 100

Steel	Normalizing temperature, °C 880	950	1050	1100	1150	1200
0·002%N	29	1000		10		5
		9				
0·0032%N	36	14·4		10·5		14
0·0047%N	18	10·3		9·6	10·7	
0·007%N	14	10				
0·010%N	17	1000		5		9
		6				
0·016%N	20M	9·5		11·1	15·2	
0·03%Al	80	14·3	17·6		14	
0·08%Al	64			13·9		14
0·1%Al	70	12·5		14·2		14·8
0·2%Al	46	9·4		18·9		29·1
0·5%Al	141	152·4		19·3	22	22
1·0%Al	28	30·6		17·9	16·7	
0·03%Si	40	15		11		14
0·3%Si	68	15·6	10·3		12·2	
0·5%Si	25	27·8	16·6		12·4	
1·0%Si	195	181	36·3		16·9	
0·016%Zr	18M	14·5		10·8	13·3	
0·048%Zr	64	14·7		16·0	22·1	
0·057%Zr	61	21		15·9	16·5	
0·02%Ti	144	221		158·0	130·4	

M = mixed grain size.

section a distinction is drawn between the effect of composition and heat-treatment on (*a*) the temperature range over which the transition from brittle to tough fracture takes place, and (*b*) the location of the transition curve with respect to the temperature axis. The former is called the 'transition range' and the latter the 'transition temperature'. The transition temperature is, strictly speaking, the highest temperature at which a 100% cleavage fracture is observed. However, for practical purposes it is customary to refer to certain nominal transition temperatures and four such criteria have been used. The 40 ft lb, 20 ft lb, and 25% fibrous fracture have commonly been adopted in published work on the properties of commercial steels and the 40% initial fibrous crack is an arbitrary value from the initial fibrous crack transition curve selected to take account of the transition range and the transition temperature.

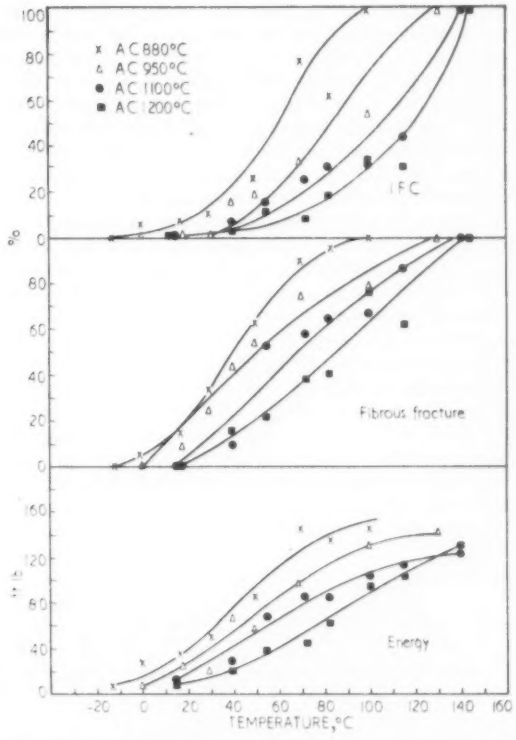
Tensile tests

Tensile tests at +18° and −192°C were carried out on specimens machined from all the steels in the normalized 880°C condition. All the tests were performed on a 35-ton Amsler hydraulic testing machine. An Amsler cold test chamber was attached to the machine for the tests at −192°C.

Specimens 0·357 in. dia. were used for the room temperature tests but to ensure that fracture occurred within the gauge length specimens of slightly smaller diameter (0·275 in.) were used for the tests at −192°C.

Microstructure

The microstructure of all steels in each heat-treatment condition has been examined in polished transverse sections of the rolled bars. When air-cooled from 880°C many of the steels exhibited a typical ferrite-pearlite structure for a 0·17%C steel but in most steels the ferrite tended to be acicular; after cooling from higher austenitizing temperatures the acicularity was more severe and thus prevented accurate ferrite grain size determination.



6 *Charpy V-notch transition curves for steel containing 0·016%N*

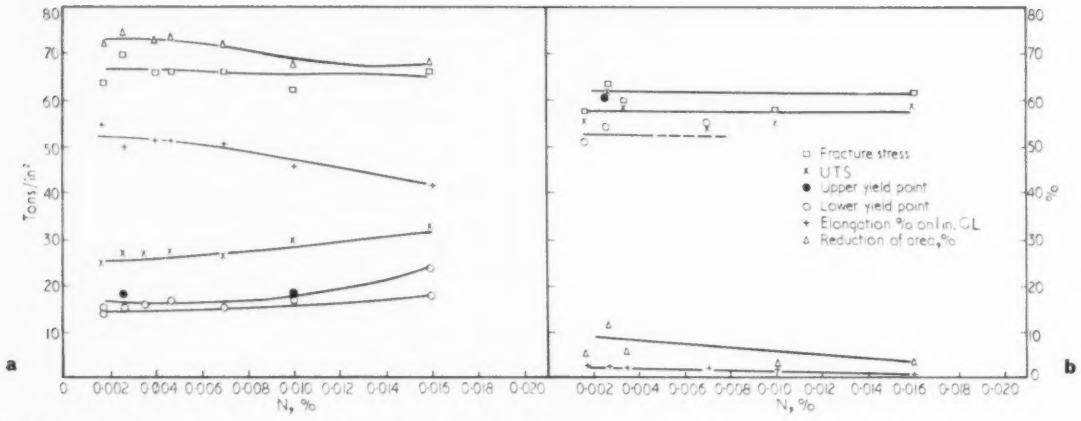
The austenitic grain sizes of all the steels after austenitizing at each temperature used were determined by the method developed by Kohn[19] and are given in Table IV.

DISCUSSION OF RESULTS

In this discussion the effects of individual elements will be considered separately. In the first place the effect of a particular element on the Charpy impact and tensile properties of steels air cooled from 880°C will be discussed; the effect of higher austenitizing temperatures on the Charpy properties will then be considered.

Effect of nitrogen

It will be observed from the six transition curves in Fig.1 that additions of nitrogen from 0·002% to 0·010% progressively broaden the transition range of steels air cooled from 880°C. The broadening of this range is presumably due to the nitrogen in solid solution or precipitated as iron nitride. Increasing the nitrogen content from 0·01% to 0·016% appears to have little further effect on the transition range which remains at about 100°C. While increasing the nitrogen content broadens the transition range the highest temperature at which a fully crystalline fracture is observed is not affected significantly by the nitrogen additions. This means that if the steels are compared on the basis of a transition temperature corresponding to the lower temperature end of the transition curve, such as 20 ft lb or 25% fibrous fracture, increasing the



3 *Effect of nitrogen on tensile properties of steels air cooled from 880°C*

above room temperature were heated in water or in an oil bath stabilized at the testing temperature. To minimize any possible ageing the specimens were kept in the oil or water for 10 min only.

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Initial fibrous crack The energy and percent fibrous fracture (or percent cleavage fracture) criteria have been employed by previous workers but these values are influenced by the resistance of the steel to the propagation of the fracture as well as its resistance to the initiation of a crack. On the assumption that the fibrous crack progresses from the base of the notch across the specimen and the length of this crack is unaffected by the resistance of the steel to the propagation of a brittle crack, the length of the initial fibrous crack was adopted as a third criterion. This was measured from the photographs of the fractured surfaces and expressed as a percentage of the original depth of the specimen below the notch (8 mm).

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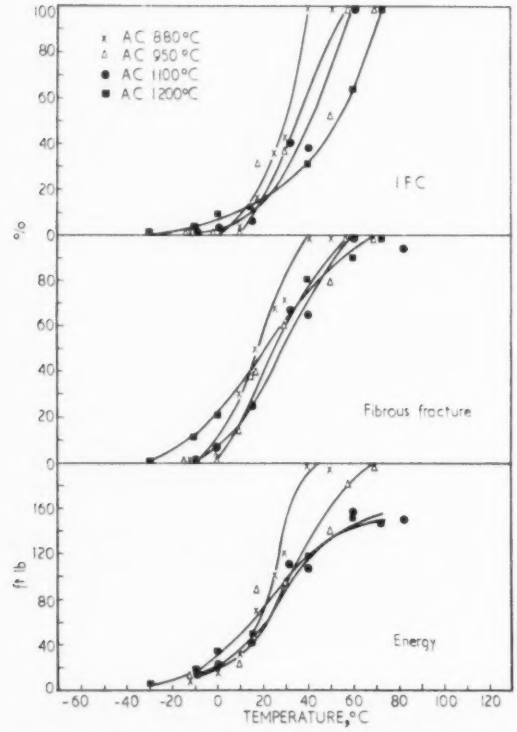
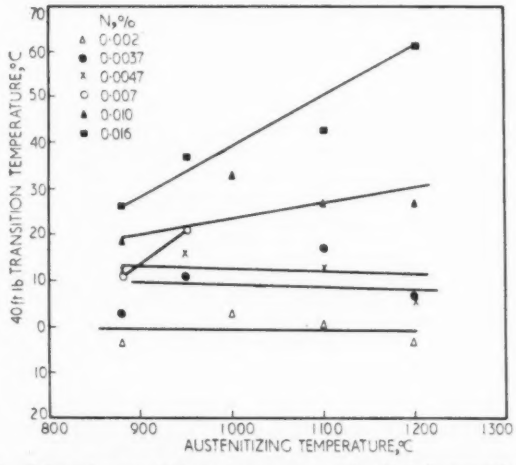


4 *Charpy V-notch transition curves for steel containing 0·0047%N*



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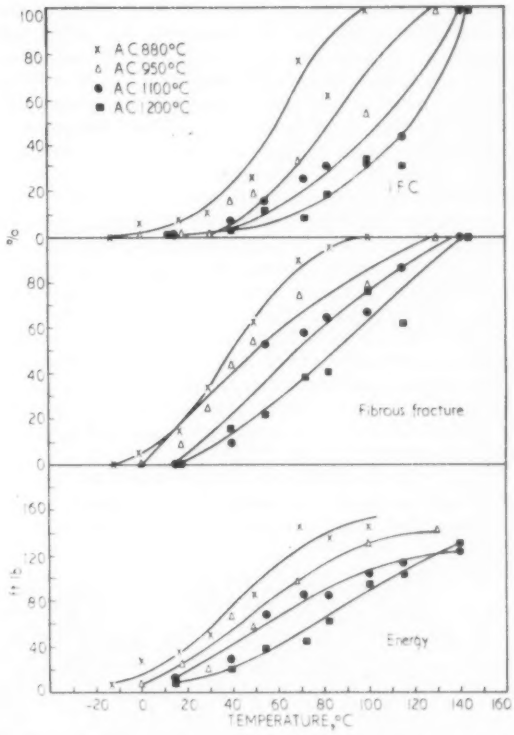

6 *Charpy* V*-notch transition curves for steel containing 0·016%N*

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Effect of nitrogen

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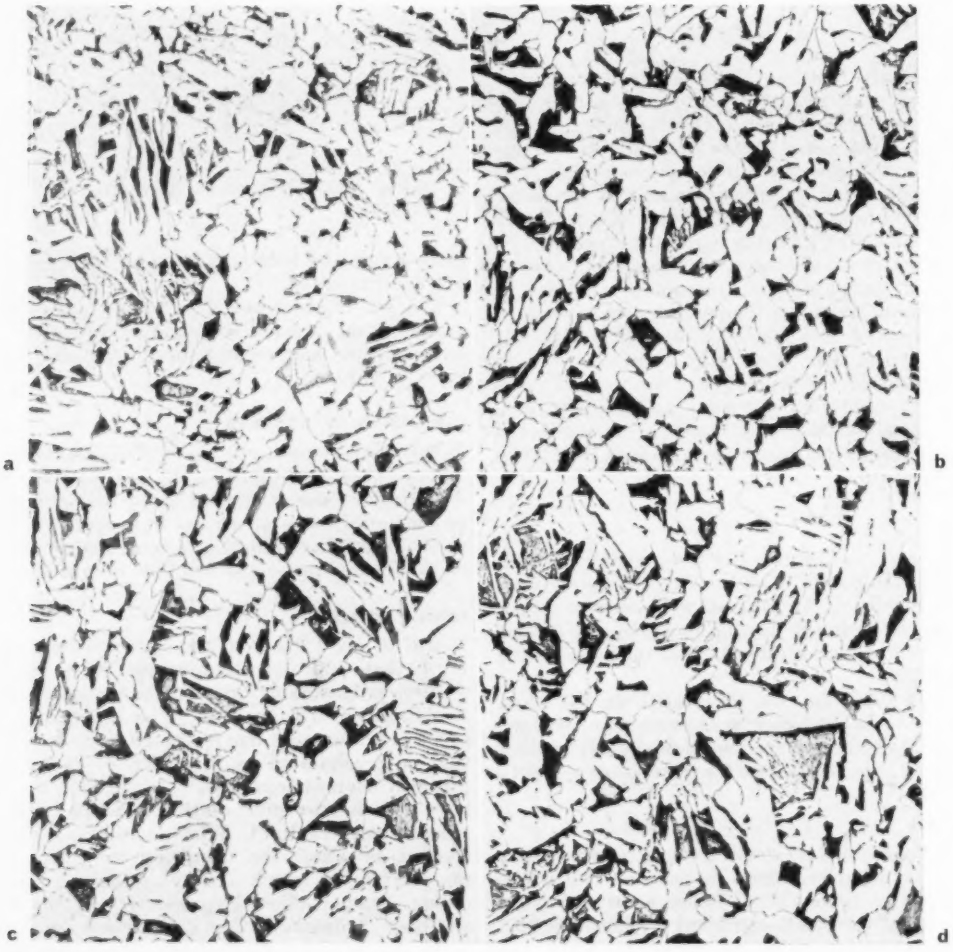
nitrogen content will appear to have little effect on the transition temperature. If, however, the steels are compared on the basis of a transition temperature such as 80 ft lb energy, 40% fibrous fracture or 40% initial fibrous fracture which represent the centre portions of the transition curves, a significant difference in transition temperature will be recorded. The effect of nitrogen content on the transition temperature as determined by these criteria is shown in Fig.2.

Additions of nitrogen up to 0·007% have no significant effect on the tensile properties at +20°C as shown in Fig.3*a* although increasing the nitrogen to 0·016% tends to raise the upper yield and tensile stresses and reduces the elongation. At −192°C (*see* Fig.3*b*) additions of nitrogen up to 0·010% have no effect on the tensile properties. At all the nitrogen levels the fracture of the tensile specimens was predominantly by cleavage; no definite yield point was detected and fracture took place without any significant prior deformation of the metal.

Raising the austenitizing temperature from 880° to 1 200°C has no significant effect on the Charpy V-notch brittle–tough transition curve for steels containing up to 0·01%N as shown in the typical group of transition curves for the steel containing 0·005% nitrogen in Fig.4. The relationship between austenitizing temperature and transition temperature is shown more clearly in Fig.5, where the 40 ft lb transition temperatures are plotted against austenitizing temperatures for each of the nitrogen bearing steels. There was a tendency for the transition temperature of the steel containing 0·016% nitrogen to rise as the austenitizing temperature was raised from 880° to 1 200°C as shown by the transition curves in Fig.6.

The microstructure of all the steels air cooled from 880°C were of the ferrite–pearlite type in which the ferrite had a slightly acicular form, due to the fairly rapid cooling rate. Raising the austenitizing temperature from 880° to 1 200°C progressively increased the acicularity but had little effect on the actual ferrite grain as shown in Fig.7. Some slight coarsening in the austenitic grain size was observed in all these steels as the austenitizing temperature was raised from 880° to 950°C but raising the austenitizing temperature further to 1 200°C caused a relatively small increase in grain size.



Air cooled from (*a*) 880°C (*b*) 950°C (*c*) 1 100°C (*d*) 1 200°C

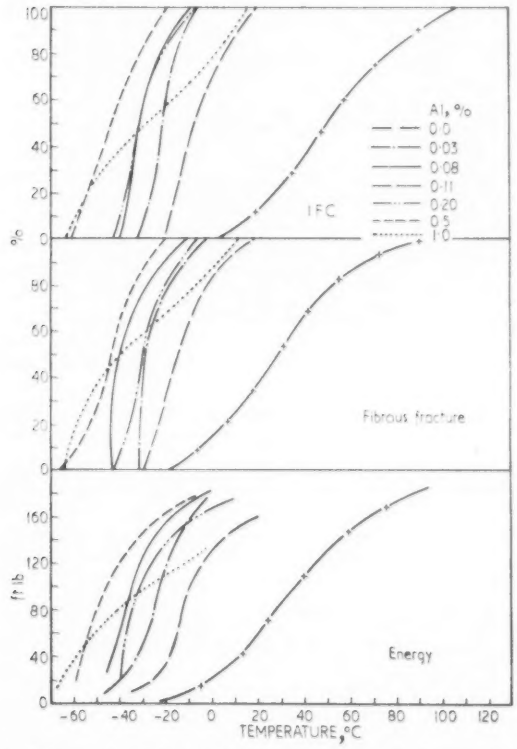
7 *Microstructures of steel containing 0·0047%N* ×200

Effect of aluminium

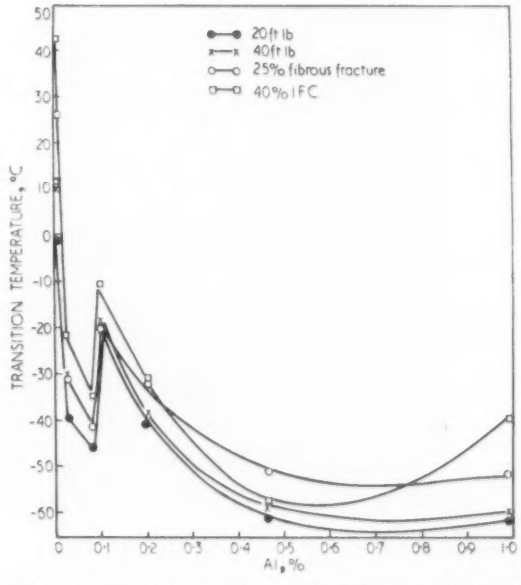
In all steels containing aluminium the transition range is of the order of 20°C. The effect of aluminium on the transition temperature is interesting in that as the aluminium is increased to 0·08% the transition temperature is progressively lowered as shown in Fig.8. The steel containing 0·1% aluminium, however, has a higher transition temperature, and as the aluminium content is increased from 0·1 to 0·5% the transition temperature is again lowered. The narrowing of the transition range is considered to be due to formation of aluminium nitride at the expense of the nitrogen in solid solution and in iron nitride. Attempts were made to determine the amounts of aluminium nitride present, but as it is not considered that the methods at present available for this determination are sufficiently reliable, the results have not, therefore, been given in this paper. The amounts of nitrogen present as aluminium nitride and as 'free nitrogen' have been calculated and are given in Appendix II. The term 'free nitrogen' refers to nitrogen present in solid solution or as an iron nitride. The lowering of the transition temperature, however, is considered to be due to the combined effect of aluminium on the ferrite grain size and the aluminium in solid solution. The effect of aluminium additions on the transition temperature is shown in Fig.9. Additions of aluminium up to 0·08% progressively lower the transition temperature and produce a fine and uniform grain size, but at 0·1% a less regular grain size is produced, as shown in Fig.10. The increase in grain size with additions of more than 0·08% aluminium is in agreement with the observations of Halley,[12] Miller,[11] and Chatterjea and Nijhawan.[13] Further increases in the aluminium content up to at least 0·5% have little effect on grain size and the lowering of the transition temperature is probably due to the aluminium in solid solution. The transition temperature for the steel containing 1·0%Al is about the same as that for the steel containing 0·5%Al. It is noted, however, that the transition curve for this steel exhibits an inflection or double transition possibly associated with some intergranular fracture.

The addition of aluminium up to 1·0% has no significant effect on the tensile properties at +20°C, although as shown in Fig.11*a* some scatter in the reduction of area and fracture stress values can be observed around the 0·1% aluminium level. At −192°C (Fig.11*b*) the addition of up to 0·08%Al progressively raises the fracture stress from 54 tons/in² to 88 tons/in² and raises the reduction of area from 1·3% to 46%. Increasing the aluminium to 0·1% lowers the fracture stress to 76 tons/in² and the reduction of area to 26%. With further increases in aluminium up to 0·5% the fracture stress rises again to 89·6 tons/in² and the reduction of area to 57%. The lowering of the fracture stress and reduction of area at 0·1%Al corresponds with the rise in Charpy V-notch transition temperature at this aluminium level.

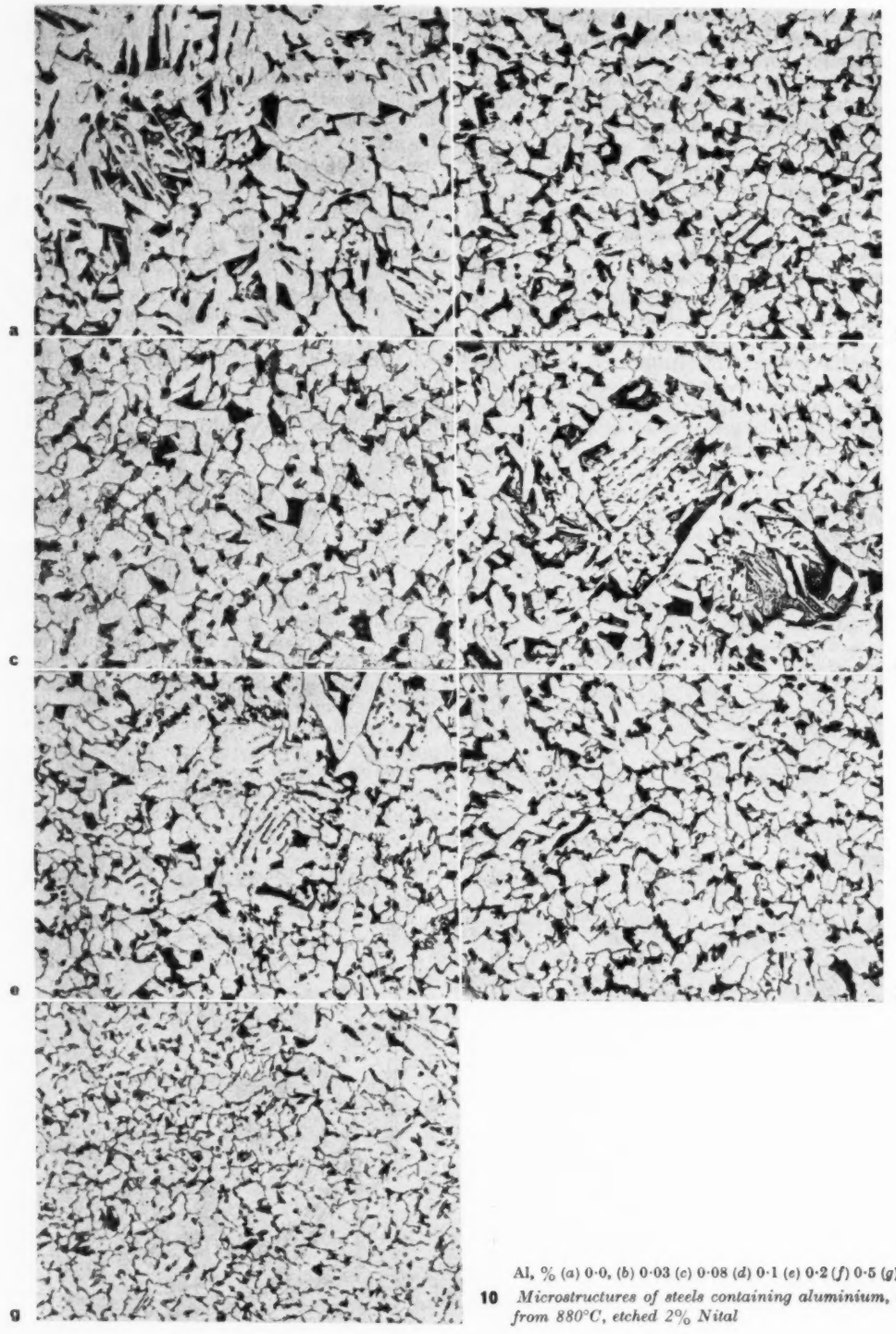
Raising the austenitizing temperature from 880° to 1050°C moved the Charpy V-notch transition curves to higher temperatures but did not significantly alter the transition range for the fine grained steel containing 0·03%, as shown in Fig.12. A similar effect was observed for steel containing 0·08%Al. These steels exhibited larger grain sizes after air cooling from the higher austenitizing temperatures. When the steels were austenitized at 1100°C or above, however, the transition temperature range was broadened from 20° to 100°C. It is possible that the cooling rate in these experiments was too rapid for all the aluminium and nitrogen in solution at these temperatures to



8 *Charpy V-notch transition curves for steel containing 0·007%N and aluminium, air cooled from 880°C*

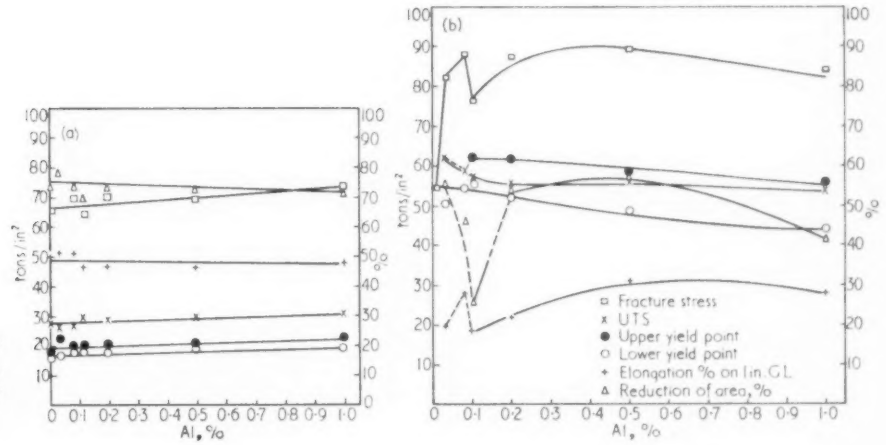


9 *Effect of aluminium on the transition temperatures of steels, air cooled from 880°C*



Al, % (*a*) 0·0, (*b*) 0·03 (*c*) 0·08 (*d*) 0·1 (*e*) 0·2 (*f*) 0·5 (*g*) 1·0

10 *Microstructures of steels containing aluminium, air cooled from 880°C, etched 2% Nital* ×200



(*a*) 20°C (*b*) −192°C

11 *Effect of aluminium on tensile properties of steels air cooled from 880°C*

recombine to form aluminium nitride; some of the nitrogen is, therefore, probably retained in solid solution and a transition range similar to that obtained in a steel containing nitrogen only is obtained.

The steels containing 0·1%, 0·2%, and 0·5%Al also showed a rise in transition temperature as the austenitizing temperature was raised but the rise was less than with the 0·03% and 0·08%Al steels. This is probably due to the smaller change in grain size produced in these steels by the higher austenitizing temperatures.

The transition temperatures for the steels air cooled from 950°C, 1100°, and 1200°C have been plotted against aluminium content in Figs.13*a–c*. The transition temperatures are lowered by aluminium additions up to 0·5% and the discontinuity at 0·1%Al occurs for all conditions of heat-treatment. The differences in transition temperature between the 0·08% and 0·1%Al steels, however, decrease as the austenitizing temperature is raised. Whereas the 0·5% and 1·0%Al steels, air cooled from 880°C, exhibited a more uniform and slightly finer grain size than the 0·1% and 0·2%Al steels, there is little difference between the structures of the steels when air cooled from higher temperatures. It appears, therefore, that while the finer grain sizes in the 0·5% and 1·0%Al steels air cooled from 880°C may have contributed to the lower transition temperatures, the aluminium in solid solution was of equal or greater importance.

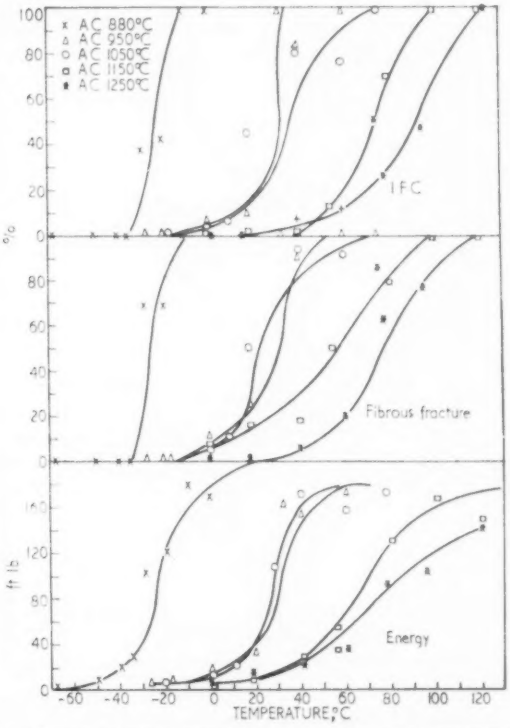
Effect of ageing on the transition characteristics of steels containing nitrogen and aluminium

The tests described above were carried out about eight weeks after heat-treatment and it was considered that some ageing might have occurred in air cooled specimens during this interval. Charpy V-notch curves obtained within 24 h of heat-treatment for the steels containing 0·007%N and 0·007%N with 0·03%Al were compared with those obtained from the same steels after about eight weeks. The transition curve for the two contents are shown in Figs.14 and 15 respectively. Transition curves for the steels austenitized at 880°C, cooled to 700°C and water quenched are also included in these figures.

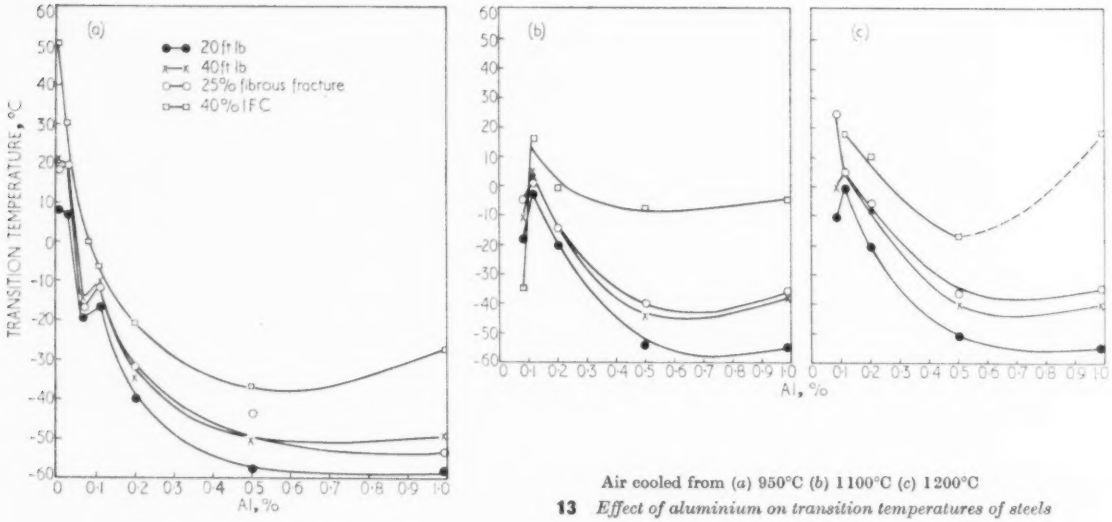
In the case of the steel containing 0·007%N, testing within 24 h of austenitizing at 880°C and water quenching from 700°C reduced the transition temperature range from about 100° to about 40°C indicating that the broad transition range of the air cooled steels may be partly due to the precipitation of iron-nitride after heat-treatment. The transition range of steels containing 0·03%Al and 0·007%N is not affected by testing within 24 h of heat-treatment or by quenching from 700°C.

Effect of silicon

Charpy V-notch transition curves for steels containing 0·007%N and between 0 and 1·0%Si air cooled from 880°C are given in Fig.16.



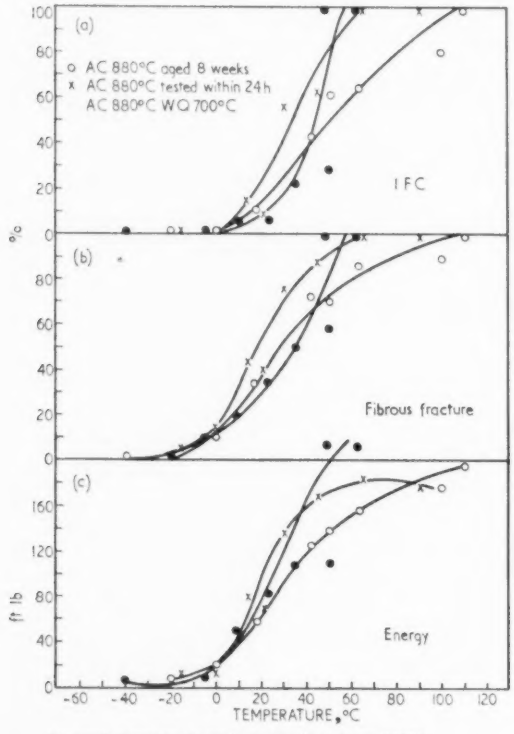
12 *Charpy V-notch transition curves for steel containing 0·007%N and 0·03%Al*



Air cooled from (*a*) 950°C (*b*) 1100°C (*c*) 1200°C

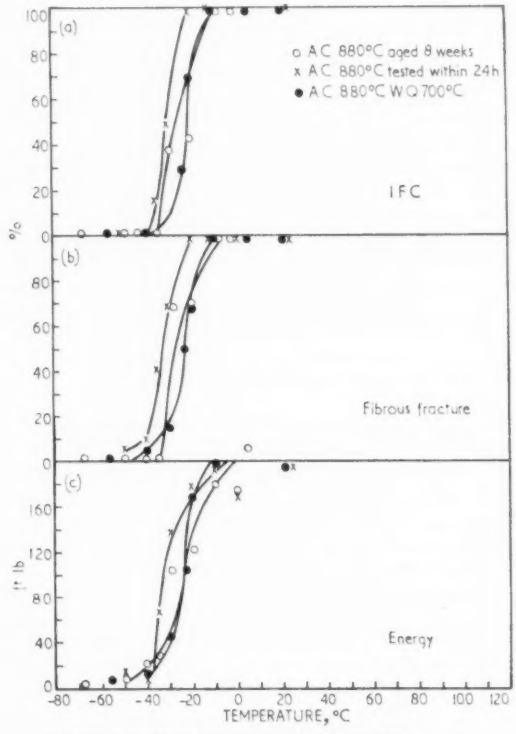
13 *Effect of aluminium on transition temperatures of steels*

The addition of up to 0·3% slightly narrows the transition temperature range and moves the whole curve to lower temperatures. Further additions of silicon up to 1·0% moved the transition curves to higher temperatures but the transition-temperature range was not altered significantly. The fibrous fracture and initial fibrous crack transition curve for the 0·3%Si steel exhibited an inflection which was not shown in the energy curve. Some intergranular fracture was observed in secondary cracks close to the main fractures and it is possible that the shape of the curve may be explained by these cracks. The nominal transition temperatures are plotted against silicon content in Fig.17 and show a minimum at 0·3%Si which is in agreement with previously published work.



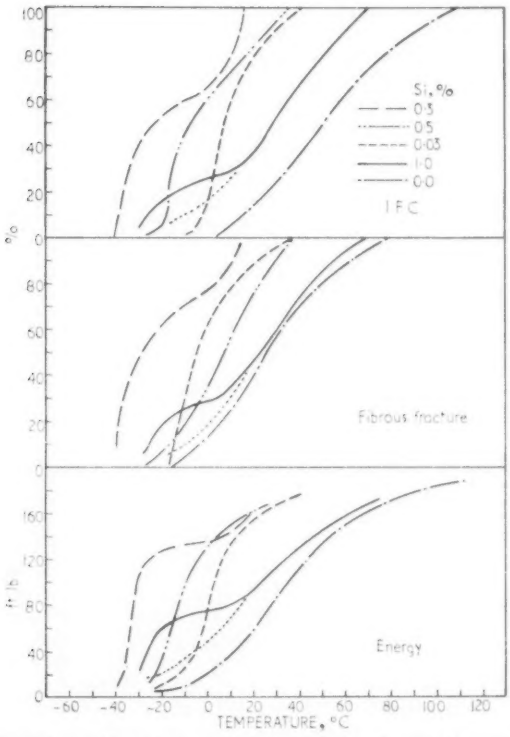
(*a*) aged 8 weeks (*c*) water quenched from 700°C

14 *Charpy V-notch transition curves for steels containing 0·007%N, air cooled from 880°C*

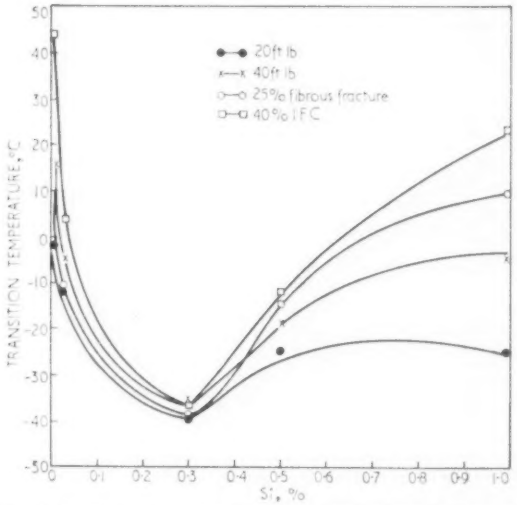


(*a*) aged 8 weeks (*c*) water quenched from 700°C

15 *Charpy V-notch transition curves for steel containing 0·007%N and 0·03%Al*



16 *Charpy V-notch transition curves for steels containing silicon, air cooled from 880°C*



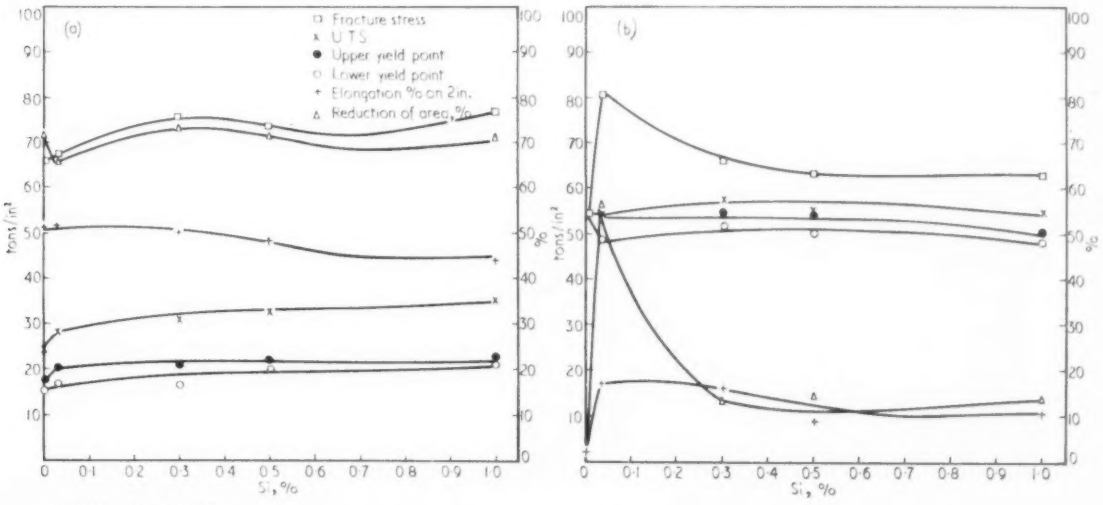
17 *Effect of silicon on the transition temperatures of steels, air cooled from 880°C*

An addition of 0·03%Si raises the yield points, ultimate tensile strength, and fracture stress and lowers the reduction of area. Further additions up to 1·0% have relatively little effect as shown in Fig.18*a*. At −192°C (Fig.18*b*), all the steels broke with a brittle fracture giving low values for elongation and reduction of area, with the exception of the steel containing 0·03%Si which had a reduction of area of 56%.

The steels exhibited well defined yield points the values of which were not significantly affected by the silicon content. With the exception of the 0·03%Si steel, the fracture stress remained fairly constant at about 63 tons/in². The higher value for the 0·03%Si steel is probably accounted for by the plastic deformation indicated by the high value for reduction of area. No maximum occurs in the fracture stress at 0·03%Si corresponding to a minimum in the transition temperature.

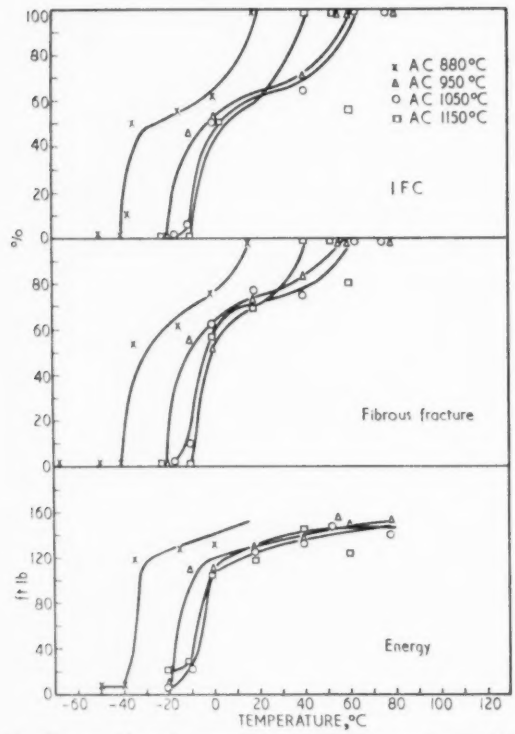
It will be seen from the transition curves for the 0·3%Si steel in Fig.19, that raising the austenitizing temperature from 880° to 1150°C has little or no effect on the overall transition range, and the inflection in the transition curve continued in the curves for the steel austenitized at temperatures up to 1150°C.

Raising the austenitizing temperature for the steels containing 0·5% and 1·0%Si had no effect on the



(*a*) 20°C (*b*) −192°C

18 *Effect of silicon on tensile properties of steels, air cooled from 880°C*



19 *Charpy V-notch transition curves for steel containing 0·007%N and 0·3%Si*

transition range but raised the nominal transition temperature; this appears to be due to a small increase in grain size.

The minimum in the transition temperature at the 0·3%Si level continued when the steels were austenitized at 950°C (Fig.20*a*) and when air cooled from 1050° or 1150°C the steel had a lower transition temperature than steels of higher silicon content (*see* Figs.20*b* and *c*). (No values for steels having silicon contents less than 0·3%, however, were obtained.)

The minimum in the transition temperature produced by the addition of about 0·3%Si confirms the work of Rinebolt and Harris[7] and of Frazier *et al.*[10] Frazier *et al.* attributed the lowering of the transition temperature to deoxidation but in view of the fact that the steels in this investigation contained less than 0·001% oxygen, this appears to be unlikely. The effect of silicon could be accounted for if silicon formed a carbide at the expense of manganese thus raising the manganese content of the ferrite. Such a substitution would be limited to a small amount of silicon and additions in excess of this limit would enter into solid solution in the ferrite and raise the transition temperature. Attempts to detect silicon in the carbide phase by analysis of carbide residues after electrolytic separation, and to find an increase in the lattice parameter of the ferrite due to manganese have, however, so far been unsuccessful.

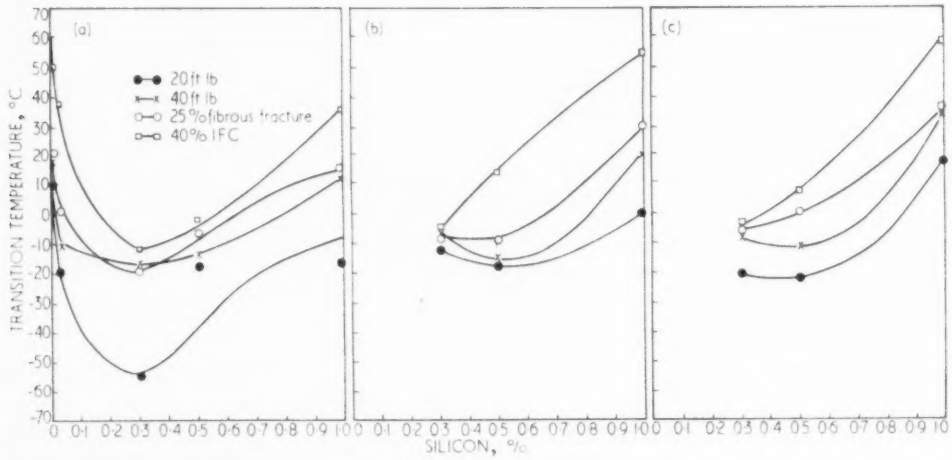
Effects of zirconium and titanium

A study of the effect of zirconium and titanium was included in this research in an attempt to assist the explanation of the effects of aluminium. Both zirconium and titanium, like aluminium, are strong nitride formers and would be expected to combine with the nitrogen and narrow the transition temperature range. On the other hand, while titanium is a strong grain refiner, zirconium does not affect the ferrite grain size.

The effects of zirconium and titanium additions on the Charpy V transition curves for steels air cooled from 880°C are shown in Fig.21. Both zirconium and titanium greatly narrowed the transition range and confirm the view that the width of this range is a function of the nitrogen present in solid solution or as iron nitride and that the addition of elements having a strong affinity for nitrogen narrow the range.

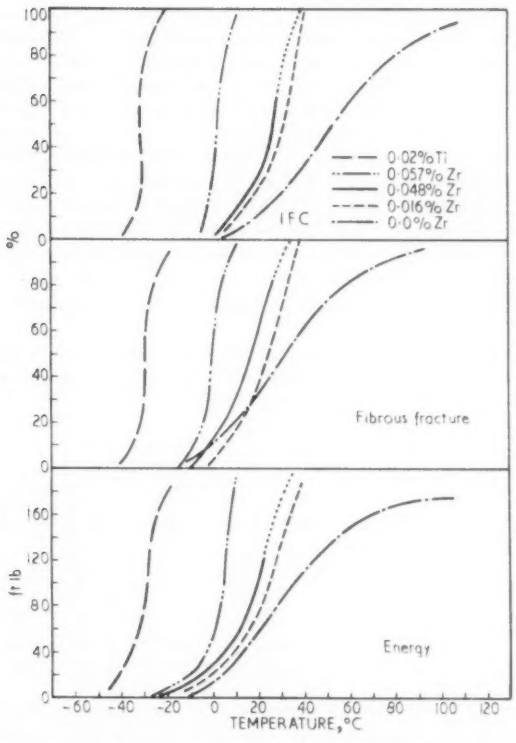
The addition of titanium moves the transition curve to lower temperatures. This steel also had a very fine grain size as shown in Fig.22 and it appears that the shifting of the transition curve is due to the refinement of the ferrite grain size.

It is observed that although the grain size of this



Air cooled from (a) 950°C (b) 1050°C (c) 1150°C

20 *Effect of silicon on transition temperatures of steels*



21 *Charpy V-notch transition curves for steel containing 0·007%N and zirconium and titanium, air cooled from 880°C*

steel is much finer than that of the steels containing 0·2% and 0·5%Al, the aluminium steels have lower transition temperatures. This suggests that aluminium in solid solution has a greater effect than grain refinement in lowering the transition temperature.

At +20°C the addition of up to 0·06%Zr has no effect on the tensile properties but at −192°C (Fig.23*b*) the fracture stress and elongation appear to exhibit a maximum at 0·047%. This is reflected in the transition temperatures of steels air cooled from 1100° and 1200°C as shown in Figs.24*a* and *b*. It is, however, not shown in the transition temperatures for steels air cooled from 880° and 950°C (Figs.24*c* and *d*).

Raising the austenitizing temperature of the zirconium steels from 880° to 1200°C has raised the transition temperature by about 20°C although the effect is less pronounced for the 0·016%Zr steel. This rise in transition temperature corresponds with a small increase in grain size.

Raising the austenitizing temperature of the titanium steels had no significant effect on the transition curves and even at 1200°C although the grain size was slightly coarser than at the lower austenitizing temperatures it was still very fine as shown in Fig.22.

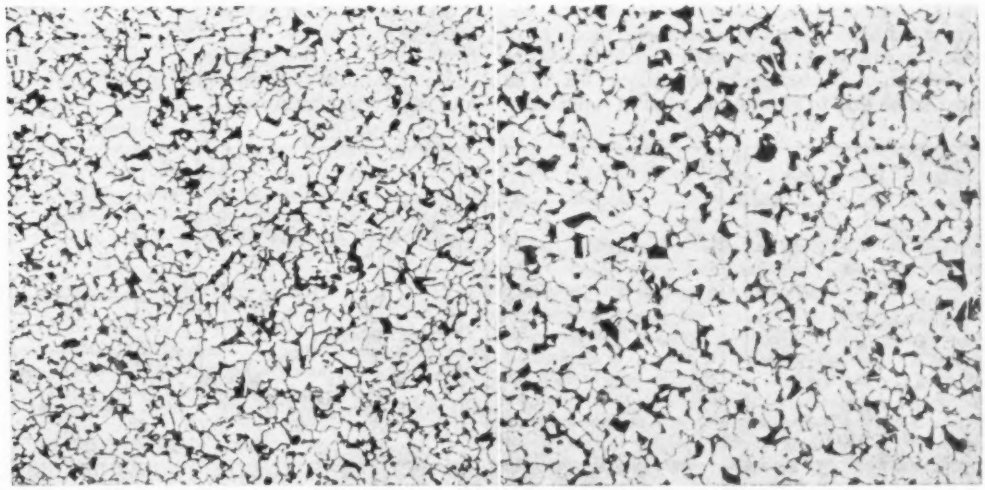
SUMMARY AND CONCLUSIONS

The results obtained from the experimental work described above on the effect of certain elements on the properties of a mild steel containing 0·17%C and 0·8%Mn made from high-purity materials may be summarized as follows:

Effect of nitrogen

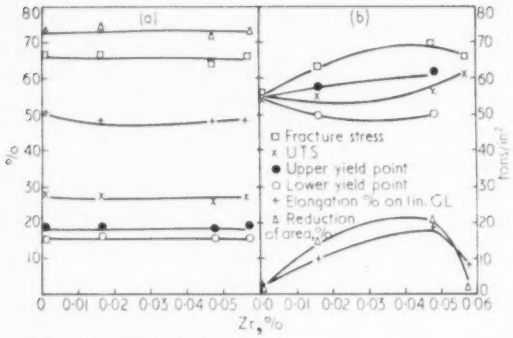
1. Raising the nitrogen content from 0·002% to 0·010% progressively broadens the Charpy V-notch transition temperature range from a few degrees to 100°C for a steel air cooled from 880°C. Increases above 0·010% appear to have no further effect on the transition range.

2. Raising the nitrogen content from 0·002% to 0·016% has no significant effect on the transition temperature (defined as the highest temperature at which a 100% cleavage fracture is observed) in steels air cooled from 880°C. Consequently nitrogen in these amounts has little effect on the 20 ft lb Charpy V-notch transition temperature. Owing to the effect of nitrogen on the transition range many nominal transition temperatures representing the middle positions of the



Air cooled from (*a*) 880°C (*b*) 1200°C

22 *Microstructures of steel containing 0·02%Ti*



(*a*) 20°C (*b*) −192°C

23 *Effect of zirconium on tensile properties of steels air cooled from 880°C*

transition curve show a rise as the nitrogen content is raised.

3. Variations in the austenitizing temperature within the range 880°C to 1200°C have no significant effect on the Charpy V-notch transition curve for steels containing up to 0·016%N.

4. Raising the nitrogen content from 0·002% to 0·016% has no significant effect on the tensile properties at +20°C or −192°C for steels air cooled from 880°C.

Effect of aluminium

5. The addition of aluminium in amounts between 0·03% and 0·5% narrows the transition range of a steel containing 0·007%N air cooled from 880°C, from 100°C to 20°C.

6. The addition of between 0·03% and 0·5%Al progressively lowers the transition temperature of a steel containing 0·007%N air cooled from 880°C. A discontinuity in this relationship occurs at about 0·1%Al. Steels containing less than this amount of aluminium exhibit a fine and uniform grain size. Steels containing 0·10% or more of aluminium exhibit a generally fine but rather irregular grain size and the discontinuity in the relationship between aluminium content and transition temperature is considered to be due to the irregular grain size of steels containing more than 0·08%Al. The lowering of the transition temperature as the aluminium content is raised above 0·1 % is considered to be due to aluminium in solid solution.

7. The relationship between aluminium content and transition temperature persists with steels austenitized within the temperature range 880°C to 1200°C although the rise in transition temperature at 0·1%Al decreases as the austenitizing temperature rises.

8. Steels containing aluminium when air cooled from 1100°C or above exhibit a wide transition temperature range of the order of that produced by steels containing no aluminium. It is considered that the steels cooled at such a rate that the aluminium nitride which dissociated at the austenitizing temperature did not re-combine.

Effect of ageing on steels containing nitrogen and aluminium

9. Steels containing 0·007%N exhibit a small increase in the transition temperature range when aged for eight weeks which is not observed when aluminium is present.

Effect of silicon

10. The addition of silicon up to 0·3% progressively lowers the transition temperature of a steel containing 0·007%N air cooled from 880°C but further increases raise the transition temperature.

11. Raising the austenitizing temperature from 880° to 1150°C raises the transition temperatures of steels containing silicon. The silicon content to produce the minimum transition temperature is not affected by variations in the austenitizing temperature within this range.

12. Steels containing additions of silicon tend to show an inflection in the transition curve which may be associated with intergranular fracture.

Effect of zirconium

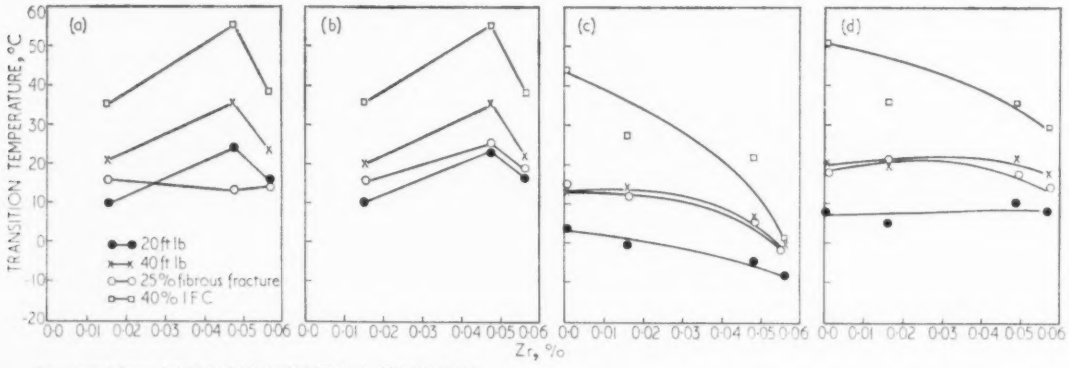
13. The addition of up to 0·06%Zr to a steel containing 0·007%N narrows the transition temperature range when air cooled from 880°C to less than 20°C, but has little effect on the transition temperature.

14. The transition temperature of steels containing zirconium is raised by up to 30°C as the austenitizing temperature is raised from 880° to 1200C° and is accompanied by a small increase in grain size.

Effect of titanium

15. The addition of 0·02%Ti narrows the transition temperature range of a steel containing 0·007%N from 100°C to a few degrees.

16. Titanium also lowers the transition temperature and refines the grain size.



Air cooled from (*a*) 1100°C (*b*) 1200°C (*c*) 880°C (*d*) 950°C

24 *Effect of zirconium on transition temperatures of steels*

17. The transition curve for the steel containing titanium is unaffected by variations in austenitizing temperature within the range 880° to 1200°C and no significant change in grain size occurs over this temperature range.

General conclusions

This work on synthetic steels made from high-purity material enables some general conclusions to be drawn on the effect of added elements and heat-treatments on the notch ductility of mild steels.

1. In the absence of strong nitride formers the transition temperature range broadens as the nitrogen content of the steel is increased up to 0·010%.

2. Additions of the strong nitride formers aluminium, titanium, and zirconium to steels containing nitrogen combine with the nitrogen and narrow the transition temperature range.

3. The transition temperature (i.e. defined as the highest temperature at which a 100% cleavage fracture occurs) is lowered by (*a*) a refinement of the ferrite grain size produced by added elements, and (*b*) the presence of aluminium in solid solution in the ferrite. The effect of aluminium in solid solution is greater than that produced by refining the ferrite grain size.

4. The transition temperature of steels can be raised by grain coarsening heat-treatments although above a certain grain size the effect is small.

5. A steel containing a nitride former will have a wide transition range if cooled sufficiently rapidly from a higher austenitizing temperature.

Evidence has also been obtained showing that steels containing 0·3% or more of silicon or 1%Al show a double transition curve and a small proportion of intergranular fracture.

ACKNOWLEDGMENTS

The authors wish to record their appreciation to members of staff in the BISRA Metallurgy (General) Division who assisted in the research. In particular they wish to acknowledge the help and encouragement given by Professor G. Wesley Austin, O.B.E., and Mr E. Ineson, the assistance given by Mr M. Ward in making the steels, by Mr K. Speight for analysing samples, and by Mr P. Taylor for his help in the mechanical testing.

Thanks are also due to the staff of the Metallurgy Division of the NPL for their help in rolling the steels.

We also wish to acknowledge the loan of an Amsler cold-test chamber used for the low temperature tensile tests from the Engineering Department of Sheffield University.

APPENDIX I

PREPARATION OF SYNTHETIC STEELS

Melting equipment

All the steels were made in the Gereatebau-Anstalt Balzers high frequency vacuum furnace VSG 25 at the BISRA Metallurgy Divisional Laboratory, Sheffield. This furnace consists of a 25 kg rammed magnesia crucible heated by induction from a water-cooled coil contained in an 18½-ft³ stainless-steel water-cooled chamber. A simple valve system allows controlled amounts of gases to be circulated through the chamber. The gases enter the chamber through a 1-in. dia. pipe directed over the crucible and leave through a pipe at the bottom of the chamber.

During the deoxidation of a melt in the crucible, hydrogen was circulated through the chamber at a rate of 20 ft³/min. On leaving the chamber the hydrogen was pumped through two consecutive copper water-cooled coils immersed in freezing mixtures of methylated spirits and solid carbon dioxide. At the beginning of the deoxidation treatment the first coil was cooled to 0°C and the second to −15°C. As the deoxidation continued and the water vapour pressure in the hydrogen decreased, the temperatures of the cooling coils were, therefore, progressively lowered maintaining a temperature difference between the coils of about 10–15°C, until temperatures of −60°C to −70°C respectively for the two coils were reached. The coils were of sufficient size to hold all the water vapour from the deoxidation of 50 lb of iron and they were drained after each melt.

Melting procedure

A high-purity Swedish iron (BISRA catalogue reference AP) was used as a base for all the 'steels'. The composition of this iron as given in the BISRA catalogue is as follows:

Chemical analysis, %

C	Mn	Si	S	P	(Total Al)*
0·017	0·006	0·001	0·007	0·002	(0·002)

* Not quoted in catalogue but determined subsequently

Spectrographic analysis

Ni, Cr, Mo, Cu, V, Tr, Co, Sn, Pb, As, B, each less than 0·01%, W 0·05%

TABLE VI Rolling programme

Pass no.	Dia. of bar
Heated to 1150°C	
1	2·593
2	2·39
3	2·218
4	2·00
5	1·825
6	1·656
7	1·625
8	1·500
Reheated to 1100°C	
9	1·277
10	1·169
11	0·982
12	0·924
Reheated to 1100°C	
13	0·686
14	0·794
15	0·737
Reheated to 1000°C	
16	0·676
17	0·625 (completely round pass)

The manganese was added in the form of electrolytic manganese of high purity and the carbon as electrode carbon of the type used for spectrographic analysis.

The Swedish iron was supplied as 4-in. square billets and in order to reduce the oxygen content of the melt to a minimum, the surfaces of some of the billets were machined before melting; others were shot blasted. The billets were placed in the magnesia crucible and heated in a vacuum; when melting began an argon atmosphere at a pressure of 1·0 cm Hg was introduced into the chamber to avoid splashing. When melting was complete the chamber was evacuated in stages over 1 h to allow the carbon in the iron to remove the oxygen by internal reduction. Hydrogen was then passed into the chamber from a cylinder, via a 'deoxo' unit for removal of the last traces of oxygen and was circulated over the melt for 3 h, at just below

atmospheric pressure. At the end of the first hour the manganese was added to the iron and the subsequent circulation of hydrogen removed the small amount of oxygen introduced with the manganese. At the end of 3 h the hydrogen pressure in the chamber was reduced to 50×10^{-3} cm and the temperature adjusted to 1600°C. In the case of the nitrogen bearing steels, controlled quantities of high-purity nitrogen gas were introduced at this stage. The pressure of nitrogen introduced was determined from thermo-dynamic data and details of the partial pressures of nitrogen employed are as follows:

Nominal N content of 'steel', %	Partial pressure of nitrogen in chamber atmospheres, %	Actual nitrogen content of 'steel', %
0·002	...	0·002
0·0035	0·0076	0·0032
0·005	0·0156	0·0047
0·007	0·0306	0·007
0·010	0·0625	0·010
0·015	0·16	0·016

The iron was held at 1600°C in contact with the nitrogen for 2 h. In the case of the 'steels' containing aluminium, silicon, or other additions, predetermined weights of these elements were added at the end of this period. The carbon was added to the crucible when all the other additions had been completed. While the carbon was dissolving an argon atmosphere was introduced into the chamber at a pressure of 5 cm Hg. As soon as the carbon had dissolved, the 'steel' was cast into a 4-in. square C.I. mould coated with zirconium silicate paste. The final weight of each ingot after cropping was about 40 lb.

The compositions of the 'steels' as obtained from analysis of samples of rolled bars are given in Tables I–III. All the elements were analysed by chemical methods. A vacuum fusion method was used for the nitrogen and oxygen determinations and a special colorimetric method developed by Scholes[20] was used for the analysis of the soluble aluminium.

Forging and rolling schedules

The cast ingots were forged to 2-in. square bars in two stages under the BISRA 200-ton forging press working to a programme so that each ingot received the same rate of working. The ingots were heated to 1200°C in a gas-fired furnace and in the first stage they were forged to 2½ in. square. After reheating to 1200°C they were forged to 2 in. square. (Owing to non-availability of the BISRA press some of the alloys were forged to the same programme on a similar industrial press.)

The forged bars were then rolled to ⅜-in. dia. at the Metallurgy Division, National Physical Laboratory, Teddington, in 17 passes with four reheatings. Details of the rolling programme are given in Table VI. All the passes except the seventeenth were nearly round and the bars were rolled twice through each, being turned through 90° after each pass.

APPENDIX II

Calculated distribution of nitrogen and aluminium

Alloy	Chemical analysis Calculations, % Al, % (free soln.)	N, %	N as AlN	AlN	'free N'	Al in solid soln.
PA	0·0015	0·002	0·00078	0·00228	0·00122	...
PM	0·0018	0·0032	0·00094	0·00274	0·00226	...
PR	...	0·0047	...	...	0·0047	...
PC	0·001	0·007	0·00052	0·00152	0·00648	...
PG	0·001	0·010	0·00052	0·00152	0·00948	...
PS	0·0014	0·016	0·000728	0·002128	0·015272	...
PD	0·03	0·007	0·007	0·0205	...	0·0165
PE	0·08	0·0078	0·0078	0·02285	...	0·0649
PN	0·111	0·0078	0·0078	0·02285	...	0·0959
PO	0·193	0·0073	0·0073	0·0214	...	0·1789
PP	0·456	0·0062	0·0062	0·018166	...	0·44034
PQ	0·995	0·0062	0·0062	0·018166	...	0·98303

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Serrations in the stress/strain curve of cold-worked 301 stainless steel at 20°K

J. F. Watson and J. L. Christian

RECENT INVESTIGATIONS of flow and fracture in various metals and alloys at 4·3°K and 20°K have described serrations (i.e. sudden discontinuous drops in load) observed in the stress/strain curve obtained during the tensile testing of these materials.[1–3] The cause of these serrations has been variously attributed to: thermal instability (i.e. plastic strain liberating enough thermal energy on certain slip planes to cause increases of temperature of a magnitude large enough to reduce the flow stress appreciably[4]; deformation twinning[5]; and burst type formation of dislocations.[6] Similar serrations and stress/strain curves have been obtained in this laboratory during the tensile testing of extra full hard 301 stainless steel sheet at 20°K, and it appears that the mechanism of deformation accounting for the observed serration in this particular alloy is the discontinuous occurrence of the austenite to martensite reaction. It is believed that the sudden shear type formation of martensite caused the drops in load due to the greater volume of the martensite phase which momentarily released the load on the specimen.

This grade of steel was cold rolled 60% and exhibited a strong directionality as measured by differences in mechanical properties between the longitudinal and transverse direction. The major alloying elements were chromium (17·64%), nickel (7·11%), and carbon (0·08%). During tensile tests of this alloy at 20°K, repeated discontinuous drops in load of about 15% were observed, with 10 to 15 drops being typical. Examination of the samples, disclosed serrations running at a 60° angle (measured from the longitudinal axis of the specimen) diagonally across the width of the sample whose appearance was not unlike Luder's bands.

Tests were run on a mechanically driven Baldwin tensile machine at a strain rate of 0·001 in./in./min. This was a 'hard' machine, and consequently the load drops were observed quite readily. The number of serrations in the sample generally corresponded to the number of drops in load, although exact correlations were impossible because many of the serrations intersected or were not clearly discernible.

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Magnetic and X-ray diffraction measurements of the reduced section of the sample before and after testing revealed that additional amounts of martensite had formed during testing. This reaction is known to be favoured by a combination of low temperature and high tensile stress.[7] Fractures occurred in a ductile, shear type manner, but the angle of fracture was inclined at 60° rather than the normal 45° from the longitudinal axis. It is postulated that this condition was caused by the formation of martensite from a highly oriented parent austenite matrix. Further circumstantial evidence was obtained from tests of cold rolled type 310 stainless steel, where no serrations and no magnetic effects were observed after identical tests at 20°K. Due to its high nickel content (20%), this steel is fully stable under all combinations of low temperature and high tensile stress.

Since these samples were very thin (0·015 in.), thermal energy evolved during straining was dissipated rapidly into the surrounding liquid hydrogen. Thus the thermal instability mechanism for serrations is believed to be a second order effect under these conditions. Neither metallographic nor X-ray diffraction techniques yielded any information relating to the occurrence of twinning during straining because of the extremely fine grain size of this steel (not resolvable at ×1000). However, this mechanism does not seem likely in a highly strained fine grained material.

Work at this laboratory has also shown that the occurrence of serrations in 300 series stainless steel sheet metal is dependent upon material variables (i.e. chemistry, amount of cold work, and thickness) and upon testing conditions (i.e. testing temperature, strain rate, and existence of stress concentrations). Further study of these variables is in progress.

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Discussion at the Autumn Meeting 1959

The Autumn General Meeting of The Iron and Steel Institute was held on Wednesday and Thursday, 2–3 December 1959 at the Great Hall, Caxton Hall, Caxton Street, London SW1. The President (Mr W. Barr O.B.E.*) was in the Chair during some of the sessions, and it was later occupied by Mr F. J. Somers. Technical sessions A and B were held at Caxton Hall, while technical sessions C–F were held at the Hoare Memorial Hall, Church House, Great Smith Street, London SW1. Part of the discussion at technical session A has already been published in the July issue of the* Journal; *the remaining discussions are given below.*

Joint discussion on forging

Part 1

This discussion was based on the following paper which was published in the October 1959 issue of the *Journal* (vol.193). Inclusive page numbers are given in parentheses.

'Operational research studies in the forging industry', by *R. H. Collcutt, J. Banbury,* and *J. V. Chelsom* (148–157).

Mr **R. H. Collcutt** (BISRA) presented the paper.

Mr **H. J. Webb** (Joseph Webb and Sons Ltd): Of the many branches of the steel industry, the heavy forgemasters' must be the most complex. Although a comparatively small industry, it manufactures a wide variety of products of many shapes and sizes and in many qualities of steel. There are 30 to 40 firms in this heavy forging industry, mostly highly individual in their outlook, some specializing in one or two products, but the great majority working in a very wide field and of what one might call a jobbing nature.

We ought, therefore, to congratulate Mr Collcutt and his colleagues on venturing into this difficult field and producing in their paper such an excellent survey of the comparative efficiencies of the forges that they have visited.

The difficulty in finding forges engaged on a similar repetition product is apparent, and the choice of four firms engaging on a simple cogging operation on tool steel ingots was undoubtedly a very good first choice, although even here the size of plant varied from a 50-cwt hammer to a 1000-ton press. I would assess a 50-cwt hammer as equal to a 250-ton press, so that we have a power ratio between the various plants of 4:1, and I would say that certainly the 1000-ton press would be far more efficiently employed on larger work, and therefore the costs from that plant which are shown in the paper are surely not representative of the average from that press.

Mr Collcutt has mentioned furnacing as highly inefficient in the industry. That is undoubtedly so, but here again, because of the wide range of steel sizes that we have to use, one cannot expect a high efficiency on one particular size of steel which is selected for examination.

The main conclusion to be drawn from the paper is surely that the essential factor is balance between heating, transporting, and forging. It is no good having a powerful high-speed press if the furnaces cannot keep up with it, and the authors have emphasized this essential conclusion in the paper. It must have been a depressing experience for the team to witness the 2250-ton press working less than 20% of its time, less than 12 min/h, and in that exerting only 20% of its power. Where press or hammer utilization is so very low, I would say that the scope for saving is so obvious that detailed investigation was not justified, and I would suggest that the BISRA Operational Research Department should consider confining their attention to examination of forges where press or hammer working is of the order of 50% of the time. I should like Mr Collcutt to comment on this suggested ratio of utilization as a criterion for starting our efforts.

Similarly, the study entitled 'Choosing between a hammer and a press' states the noteworthy fact that the hammer worked only 35% of the working shift; handling time was additional: the figure is not given directly in the paper, but it appears to be about 15%. Surely, therefore, such utilization factors should be studied first, before considering the purchase of a press, especially if the proposed press shows little gain in forging time. (In this section of the paper we are not told the size of hammer or the size of the proposed press. Whether they were in line from the power point of view one does not know.)

One complicating factor which I do not see mentioned in the paper is the question of quality of product. In connexion with quality, I am not considering the metallurgical aspect, but putting it under two headings: first, the physical soundness of the forging, which usually boils down to the amount of work put into the ingot: generally the reduction in the ingot cross-section; and second, the dimensional accuracy and finish of the forging. This second quality factor varies with the policy of the firm, and often depends on whether the forgings are dispatched to the firm's machine shop or to the customer's machine shop. Both these quality factors entail extra forging time, possibly of the order of 25% for the finishing alone. I should like to hear Mr Collcutt's comments on this, and to ask whether he has in any way taken this factor into account during the research.

At the BISRA meeting on forgings, which was held in Bournemouth in November 1959, Mr Benson and Mr Hunt gave a paper on forge tooling. This is yet another factor in production which can make a great difference to forging time.

I was interested to hear Mr Collcutt's remarks on production scheduling, and if time permits I should like him to enlarge a little on this. It is a difficult thing to control, in present trade conditions, because of our lack of long runs and owing to the low order book ahead of us.

Mr **N. Moore** (Jessop–Saville Ltd): Open-die forging is the oldest and most primitive method of shaping metal. It exists to make the odd shapes that are required in small numbers by the manufacturing engineer. Because it must cater for every whim of the customer, the machinery designer, the jobbing forge must be a very flexible plant capable of producing as

large a range of forgings as possible. This means that in general terms the process is crude and wasteful; no single forging is made as cheaply and efficiently as it would be if the shop had been laid out to specialize in its manufacture.

The paper shows us how the efficiency of an open-die forging shop can be improved, and I should like to comment only on one point. By and large the paper states that efficiency can be improved only under conditions in which output is increased, which in turn demands an increased order book. I do not believe that the order book will be improved by a greater customer demand, although there are ways in which the order load on a given shop can be increased. The demand for open-die forging does not vary proportionately to the general demand for steel. An expanding steel market does not lead to much greater activity in the jobbing forge shop, for as the batch number of a given forging increases, so does it become possible to manufacture in the specialized shops, such as the die forges, ring mills, and tube mills.

There is such a difference in price between an open-die forging and a machine-made forging that no conceivable increase in efficiency could stimulate demand for the products of the jobbing forge purely by offering a cost advantage to the customer.

I recommend this paper to the forging user to remind him how expensive an article the oddment forging is, and how difficult it is to cheapen that product.

Mr **H. U. Hattersley** (Steel, Peech and Tozer): Forgemasters are really grateful to BISRA for the liberal help and advice they are receiving and for undertaking fundamental research in a matter which has had little or no such research for years. During the war there were some drastic and radical improvements in forging methods and outputs.

In the matter of utilization, BISRA studied two forges, one of which was ours, and we had rather a staggering utilization of 75% on our press. It was due to our having an order book which permitted that sort of thing. With present competition for trade, our utilization now is nothing like that figure.

In the matter of manipulators, a few years ago we put in an axle-forging plant which was completely mechanical. The axles were made by manipulator. In the first stages the manipulator had to fix the bloom and take the axle away. We revised it so that the blooms were delivered under the hammer and could be gripped immediately by the manipulator and the axles taken away from under the hammer, so that the only travel the manipulator made was a few feet. In a plant of that scope there appears to be no need for electrical long travel. A hydraulic cylinder which actuated the long travel movement of the manipulator over those few feet would, under suitable circumstances, give an adequate return.

Mr **D. C. Harries** (English Steel Rolling Mills Corporation Ltd): The operational study on typical forges comprising both press and hammers, carried out by the authors, emphasizes in a most revealing way, the influence of mechanization and programming on the utilization of the forging unit.

The forging manipulator figures very prominently in this and there is no doubt that much scientific thought is being given to synchronizing its performance with that of the forging unit itself. Press design and performance have improved greatly in recent years and it is obvious that the manipulator must match these if the advantages are not to be lost.

The high proportion of cost represented by heating, especially inefficient heating, is also justifiably emphasized. On this point it is appropriate perhaps, to refer to the rapid rate of heating, by comparison with older traditional ideas, which are possible, provided this is symmetrically applied in well designed furnaces.

Not only heating, but steam for operating the hammers can be a very expensive commodity owing to leakages in the system and the natural difficulties in turning on and shutting off at will. Compressed air is often a more convenient and economical medium to use from these points of view.

Authors' reply

Mr **J. Banbury:** Mr Webb commented on the wide variation in size of the forges we studied in the early survey. This variation was intentional, as he appreciates, because, above all, we wanted to see what effects differences in the provision of plant and different forms of organization had upon the criterion which we were using in measuring efficiency, i.e. cost of the end product. Again, so far as the utilization of the press or hammer was concerned, we wanted to cover a wide range of utilization because we wanted to be able to predict what would be the effects of making changes, e.g. to visualize what sorts of changes could be made to increase utilization and be justified by reductions in cost or increases in total profit.

So far as the quality of the end product was concerned, we appreciated that this was a variable which we should include. However, we could not ask for measurements of quality to be made, as this would require some sort of metallurgical studies; and we therefore had to assume that with this particular product, which is fairly closely controlled, at least metallurgically, the quality was sufficiently constant for our purposes. It may be a doubtful assumption.

As for dimensional accuracy, we dealt with this as best we could on the spot, but current work in the department is aimed at a much more detailed study of the value of dimensional accuracy in forging. The aim is to minimize the cost of reaching the required tolerances in the machined end product, where one is trying to minimize the sum of the costs of forging and machining, and hence to state to what tolerances one should forge.

On the question of tooling, we have not tried to control this in the studies carried out so far, and slight differences in tooling no doubt account for some of the variability that we found between forges. However, we never encountered anything other than flat tools. We have taken the tool dimensions normally used in each case and also the usual methods of parting off.

Production scheduling has been included in our current work on forging. What sorts of improvements, in terms of reducing costs, can be effected by more rigorous methods of scheduling production through the shop? Such methods are not easily implemented effectively when order books are short, but even in these circumstances, where one can look ahead a fortnight, a month, or two months, something can be done to put the jobs into the best sequence for minimizing cost in the forge, bearing in mind the commitments undertaken by the sales organization.

Mr Moore commented on the need for flexibility in the equipment provided because of the very wide range of products to be dealt with in jobbing forging, both as to size and shape, and he concluded that the forge would therefore inherently not necessarily be the best for any particular product. I quite agree. What we hope to produce eventually is a specification for the most suitable plant and organization for a given pattern of end products so far as shape and size are concerned, in different quantities, in terms of the most suitable equipment for dealing with that degree of variation of forging complexity. Whether that will ever be possible I do not know.

I should like to comment on the ways of increasing efficiency. Mr Moore had understood that we had assumed that this could only be achieved by increasing output. However, in the hammer *v.* press study we first considered if the change would pay with the existing market: that was one of the prerequisites. Similarly, in the study of the effects of different methods of manipulating and transporting, we considered two sets of conditions: first, what sorts of changes one could afford to make, i.e. one could justify so far as cost was concerned, with the existing market; and second, what sorts of changes one could afford to make if one could increase the market to take up the full capacity of the shop.

We are very glad to have Mr Hattersley's support based upon his experience of his axle forge, for our general conclusions about the need to separate the transporting function from the manipulating function, and the effect of this upon press or hammer utilization.

Mr Harries was concerned with heating costs and the design of furnaces: how could one design furnaces for dealing with variable products? That point may be dealt with to some extent in one of the papers to be presented later at this meeting, but it seems that this also emphasizes the need for effective production scheduling in order to minimize total costs.

Mr Webb asked for details of the size of the press and the size of the hammer in the study. We regret that we cannot give this information if we are to preserve the anonymity of the company concerned. We are extremely grateful for their permission to publish the bulk of this particular study.

Part 2

This discussion was based on the following papers (dates of publication in the *Journal* are given in parentheses):

'Theoretical analysis of bloom and billet forging', by *J. G. Wistreich* and *A. Shutt* (1959, **193,** Oct., 163–176).

'Spread and elongation in flat tool forging', by *A. Tomlinson* and *J. D. Stringer* (1959, **193,** Oct., 157–162).

Dr **J. G. Wistreich** (BISRA) presented the first paper and Mr **A. Tomlinson** (BISRA) the second.

Mr **R. C. Benson** (Darlington Forge): The authors have started to tell us what to do with our fully automatic forges, and how we shall use them when we get them. That is obviously wise as a long-term policy, but it is quite uncertain when we shall get these automatic forges, especially for medium and large works. One wonders whether it will be in the lifetime of some of us who are present. At present, improvements in handling methods and in press control are visible only slowly and in small quantity.

A point that might be borne in mind is that when we do get the automatic forging plant, we can be quite sure that the forgings which such plant will have to make will still be like those made now, widely varying in size and shape. In other words, a forging is usually something with knobs on it. If the authors intend to extend this theoretical analysis to cover such a wide range of products and shapes, it is perhaps as well that the fully automatic forge is not imminent, because they will certainly have a big job on their hands. In the case of an especially awkward forging, with which I am familiar, I would certainly like to see an automatic forge making a large rudder stock with a preset programme.

To come to rather more practical points, in Dr Wistreich's curves showing the speed it is not clear whether the speed of the press was taken as constant during penetration. This makes a big difference. If it is a direct acting press, the speed is constant during the penetration stroke, whereas if it is an accumulator or intensifier press, the speed of penetration will vary with the resistance; so that there should be two different curves for the two different types of press. Dr Wistreich must have made one assumption or the other, and it would be interesting to know which.

My second point concerns the ever-present question in forging which the authors have described as side spread but which the forgeman describes much more appropriately as bellying or bulge. The side spread given, denoted by w_1/w_0 in the paper by Tomlinson and Stringer, is only the mean side spread. It is what has been lost in elongation and gone into the width. This bulge unfortunately, as shown in Fig.1, has a wavy surface, and after raising the press and turning the forging through 90°, one has to lift the press high enough to clear those waves. They have also to be squeezed down again during the next pass before getting on to the solid, as it were, and starting the dimorphous deformation again; therefore unless the maximum of those bulges has been assessed, the theory is lacking. With a constant speed press, unfortunately it takes just as long to squeeze off an inch of local projection as it does to squeeze an inch into the solid mass.

My next point concerns tool shape. The edges of the tools must be rounded or bevelled, or have a combination of the two. What limits the squeeze is the depth that the curved edge of the tool can be sunk into the surface of the steel. With, for example, a true radial fillet, say, a 3-in. radius on the side of the tool, it would be very rash to sink that tool 3 in. into the surface of the steel; almost certainly laps would result.

As a rough rule, it is wise not to go beyond the place where the original surface impinges on the radius at an angle of 45°. The curve is then at the side of the impression which terminates at 45° to the horizontal, and this should press in again at the next bite, without overlapping. With specially difficult steels, however, one must keep below even that. There are some steels, particularly the heat-resisting steels, the surface of which can hardly be forged without breaking, and it is necessary to have an even more liberal ratio of radius on the edge of the tool to depth of penetration.

Tools with a bevelled edge have been used extensively in the past. They enable one to achieve the heavy squeeze which gives rapid forging down. The difficulty is that for any given width of flat face, the bevel at either side of it (which is conveniently something like 30° to the horizontal) must not be wider than the flat face.

Perhaps it can be visualized. Imagine, for example, a tool with a flat face 6 in. wide and at either side a 30° bevel 6 in. wide, the total width of the tool (I must ask Mr Tomlinson to call it width and not length) being 18 in. After a squeeze, there is 6 in. of flat surface and 6 in. of bevelled surface joining the squeezed surface to the unsqueezed surface. For the next squeeze, it is not possible to present the flat part of the tool to the bevelled part of the impression. It must be taken up on to the original flat surface of the billet. Therefore, if the bevel is wider than the flat, it will not be possible finally to produce a flat surface by overlapping impressions.

Therefore, the only way to use a bevelled-edge tool effectively is to have a very wide tool with very wide bevels and again a very wide face in the middle. In most practical work such a tool is no use for finishing, because one cannot get up against collars or setdowns. Instantaneous tool changing is essential. This is a development required in the future, so that the forging can be finished using tools with a substantially flat face, so as to make the changes in section in the forging.

Turning now to the power of press, this must always be a compromise, because in our forging work large and small work has to be done on the same press and also large and small parts of the same forging; one is in fact confronted with the variability factor. In order that the large press shall not be too slow or too expensive, a good deal of thought has been given to two-power or three-power presses, the idea being in some cases to have a three-cylinder press and to use one centre cylinder, two side cylinders, or all three according to the pressure required and the work being done.

Another factor is the power of presses. A press is designed to give the daylight required by the maximum size of ingot, which it is designed to forge. Then, it may be noticed that large rings and discs which are bigger than the ingot require more daylight, therefore the press will be made that much larger. Next, it may be pointed out that larger ingots can be forged. And so it goes on, larger and larger ingots being forged. I do not know what we can do about this except by emphasizing that an overpowered press is definitely the right thing. At Darlington, we have a 2000-ton press which does the work normally done by 1000- or 1500-ton presses so that it is really an overpowered press. It is an old, intensified press, but on the lighter end of its work the extra power makes it an unusually speedy unit.

On forging factors: in connexion with the section of the second paper dealing with spread, I should like to ask for a further study in this direction to enable us to assess the actual

mechanical work done on the forging and its probable effect on physical properties. Mechanical work applied to the forging improves its structure by plastic flow; therefore it is necessary to measure the amount of such flow.

In many cases, forging reductions are specified, but normally only the final and starting dimensions are specified: e.g. the cross-sectional area of the ingot shall be three times the cross-sectional area of the forging. This means that the L_1/L_0 finally is 3. It does not take any account of what happens in between. This forging reduction has for many years been known in the trade by the simple term 'forging factor' denoted by the symbol *ff*.

The point of the three-figure formulae described by Mr Tomlinson is that the forging factor vertically on any one squeeze is equal to the product of the forging factor transversely and the forging factor longitudinally. In other words, if there is a small unit cube inside the steel, the height of the cube is reduced in a certain proportion and the other two dimensions in normal squeezing are increased, one being the length and the other the width. Subsequently the width is brought back again. The actual mechanical work which is done on that piece of steel in producing the final shape is the total summation of all these flows, measured by the product of the successive forging factors.

Is it possible to assess the extra work done by forging under conditions where there is a large amount of side spread? The material which has spread sideways has to be forged back again to produce the final shape, so that a unit cube within the material has been subjected to a greater forging factor than if the final shape had been produced with less spread.

We want next to know what has happened on the surface and at the centre. Can the authors give a figure for the amount of *eventual total deformation* that has taken place in the centre of the ingot and at the surface? It would be very valuable information and customers might begin to specify it.

I noticed that one of the speakers used the expression 'open dies' for 'flat tools'. This is US practice but I would suggest that the UK practice is better: the term open die might with advantage be used to describe just that, i.e. a die which is open at one side, as opposed to the dies used in drop-forging which contain the forging completely. It may be worthwhile to explain the term 'open die' more completely by a simple example. When a shaft is inserted in a plain bolster with a portion projecting and this portion is flattened, a flange is formed with a rounded edge. If the bolster has a recess in its top face, the projecting part of the shaft fills this recess when pressed into it by an ordinary flat top forging tool, so that the edge is prevented from bulging and is kept concentric with the shaft. As we often have not enough power to fill such a recess with one squeeze the die or bolster is often moved under the press so that only part of the flange is covered at a time, but the object is generally to produce a flat face on the top, or *open*, side of the die.

Mr **H. U. Hattersley** (Steel, Peech and Tozer): I can only speak on behalf of the heavy forging industry. We are more concerned with the forging of octagonal ingots than squares. The various advocates of flat tools all have their different views, but normally we try to keep away from squares. If you make a square into a round, it is easy to chill the corner, which can have a detrimental effect on the surface finish.

Secondly, we are much more concerned with what square we can get out of a particular job, and more particularly that it will have square corners. That is tied up very much with the heavy draughting. Dr Wistreich, broadly speaking, advocates heavy draughting, but if one has a heavy draught there will be a tremendous belly and when one turns that belly up and comes down, the corners will not come back to where they were. With particularly heavy draughting, if one looks at the bar in plan, it is a series of scallets. Therefore, one does not have a straight swelled edge all the way down; that comes later.

The paper by Green and Stringer postulates only the press driver who will drive the whole equipment. I know of nothing more difficult to make than a good square, and we still have to have in our forge what we have christened the 'back tonger', whose purpose is to watch the bar at every squeeze and to give a signal to bring the piece up, so that it will be a true square at the end and will not be twisted from one end to the other. There is need of a man for that purpose. That means that the detent referred to in the experimental regulator will not usually be used regularly.

No reference has been made to differential spreads or different spreads for different steels. In the rolling trade, I have heard of people throwing a bar half-rolled on to the floor and letting it cool for several minutes, and afterwards they can get tremendous spreads.

Mr **D. C. Harries** (English Steel Rolling Mills Corporation Ltd): The subject of optimum conditions of forging squeeze, feed, and number of passes in relation to output recalls to mind an earlier paper by Cook, which considered these same factors in relation to internal strains within the material being forged.

The authors have explained the principles involved in assessing the best combinations of these variables from both metallurgical and production points of view. Having established this it would seem that the problem remaining is to ensure that such conditions are regularly repeated.

It is only too true that difficulties encountered during operations subsequent to forging, e.g. heat-treatment and machining, are the direct result of uneven reheating and irregular forging technique, the effects of which are often difficult and sometimes impossible to eradicate.

Mr **Benson:** Dr Wistreich said that the manipulators that we have at present will not forge both inwards and outwards, or backwards and forwards. I can assure him that they can and do. The usual reason for forging backwards or away from the press is that the steel at the moment of deformation is plastic, and what is sticking over the anvil droops. The piece between the anvil and the manipulator is upheld by the manipulator. Therefore, the free end piece that bends down is straightened during the subsequent squeeze. That is the only reason that forgings are brought in again for the next pass.

When working quickly on special steels with small reductions, it is our practice to forge backwards and forwards, with the simple object, as Dr Wistreich rightly pointed out, of saving time. When forging 'difficult' steels, a heavy reduction cannot be made and the bending problem does not arise.

When slabs are cogged under a hammer in flat tools, we had in the past, and indeed still have, a man at each side of the hammer passing the slab backwards and forwards completely through the hammer by means of tongs which grip the side of the slab. These are known as nobbling tongs. If a manipulator at each side had similar tongs in place of the jaws which enclose the slab completely the same procedure could be followed.

Authors' replies

Mr **A. Shutt:** Mr Benson said that a major objection to this theory was the fact that forgings have knobs on. I know that most forgings do have knobs on, but in the stage between an ingot and a forging with knobs on, the forging does not have knobs on; this stage is generally called cogging. It seems to me from my small experience of watching forges that cogging takes quite a lot of time. Let us say that it takes only 30% of the total. If the cogging time can be cut down by half, 15% of the total forging time is saved; and I understand that if one can save 1% of the time or cost in the steel industry, then the saving is worthwhile. In addition to square cogging schedules we have schedules which will forge rolls with a barrel, a journal, another barrel, and another journal on the other end.

We took the speed of the press as being constant, because the press with which we have to work, a 200-ton press, has a constant penetration speed, and consequently we did not investigate variable speed presses. The fact that the penetration speed is not constant does not alter the principles of the analysis at all. It is quite easy to take the average speed and

then apply the same analysis: a differential equation could be used instead of an ordinary equation in the middle of the theory, thus taking account of the variable speeds.

Next, the question of bellied surfaces. Obviously we cannot at the moment take account of the size and shape of the bellying. This is a very difficult problem in the theory of plastic flow which no one has yet managed to solve. Empirically, Mr Tomlinson's experiments in his paper show that if one forges a bar on one pair of faces and then turns it up and forges it on the bellied surfaces, the same spread law holds for the two passes; therefore as long as we confine ourselves to average dimensions, we need not take account of bellying in our theoretical analysis, although obviously it is an important practical point. We are in the process of developing an electronic sensing device which will be able to take care of bellying.

I am sorry to have to disagree with Mr Benson about tool shape. We did some experiments recently on the incidence of lapping. The first thing our experiments showed was that whatever the shape of the edge of the tool, whether square, radiused, bevelled, or with just the corner chamfered off, if the tool is sufficiently wide as much reduction as desired can be taken, without any resultant lapping. In our laboratory we have taken reductions as large as 72% on raw ingots without producing laps. We have been able to show that the length of contact between tool and stock is much more important than reduction in determining whether or not laps occur.

In the course of our experiments, we have compared the number of laps produced by square-edged tools, by tools with radiused edges, by tools with quite large bevels, and tools with the corners cut off. Obviously, of all these the tools with the square edges produce the most laps. If, however, the bite ratio is greater than 0·5, i.e. if the width of the tool is more than half the width of the bar, one can take as big a reduction as desired on a square bar without producing lapping, even with square-edged tools.

Mr Benson said that with a radiused edge, one could only press in until the surface of the bar was at 45° to the radius. We have shown in our experiments that a 60° slope on the edge is at least as efficient in preventing laps as any radius that can be put on the edge of the tool.

A point was raised about the tool with the 30° bevel. Mr Benson said that the sloping bit could only be as long as the flat part. This holds only if a 7-in. penetration is taken at every bite.

I agree with him about the necessity for quick tool changing. If most use is to be gained from an automatic press, then it must be possible to change tools quickly, because one set of tools is not enough for every requirement. It is no good using our theory, or future developments of it, to do a job in 3½ min if two tools must be used in that time and it takes 2 h to change them.

Concerning the power of presses and the fact that compromise is necessary, we are working on a method of evolving the optimum compromise when one has a lot of different things to do on one press. At the moment, I have taken half a dozen different sizes, starting from a 24-in. ingot going down to a 6-in. bar, and from a 10-in. ingot down to a 5-in. bar; so that with about half a dozen of these varying sizes we can work out what are the optimum press sizes on the basis of the theory outlined by Dr Wistreich.

About press size, we say that if advantage is to be taken of recent developments, the same physical size of the press, i.e. the same tool size, the same daylight, etc. is desirable but with roughly double the power. I do not know what that implies. Perhaps it means having four pumps instead of two and doubling the size of the columns. It certainly does not mean doubling the whole of the press, with the vast increase in capital outlay that would entail.

The question of the mechanical work which is put into the ingot and how it is distributed from a theoretical point of view is a difficult problem. Cook, however, whose work has been referred to today, did some work on this in the paper referred to by Mr Harries. He went some way towards showing how the stresses and strains are distributed inside an ingot when it is forged.

In reply to Mr Hattersley, the analysis which has been presented today is not restricted to squares. It can be used for all sizes of rectangles from squares to wide, thin slabs or any forging which may be built from such shapes. I do not know whether that increases its use very much.

Our rolling section is doing work on the spread of different sorts of steel, and so far it has not found any difference in the spread. As far as can be told, all steels spread the same if they are given the same surface; but the work is not finished yet.

Mr **J. D. Stringer:** One point arises concerning the mechanical working of the forging. The work by Mr Cook, to which reference has been made, gives us a lead on this problem. It is not true to say that a small unit cube inside the steel undergoes the deformation which one would calculate from our paper. Each unit cube there is undergoing a different amount of deformation, and a more complicated summation than Mr Benson suggested would, therefore, be necessary.

There is, however, the point that a forging is only as strong as its weakest link. If we have cavities inside a forging, the fact that we have worked a lot of the body of the material but left the cavity untouched becomes the most important factor. My feeling is that we should study the deformation of forgings from the point of view of cavities.

My second point concerns the question of the different spreading of different materials. Figure 5 in Appendix II has been made up from some data available from the USSR, tests having been done on steel and duralumin. This figure shows that in this particular case, where blocks were being deformed, the tests on duralumin and steel are very close and lie almost on the same line.

Part 3

This discussion was based on the following paper which was published in the November 1959 issue of the *Journal* (vol.193). Inclusive page nos. are given in parentheses.

'Experimental semi-automatic forge', by *J. I. T. Green* and *J. D. Stringer* (231–237).

Mr **J. D. Stringer** (BISRA) presented the paper.

Mr **F. H. Towler** (Towler Bros (Patents) Ltd): Many of us will have seen a 500-ton press at Walter Somers doing 140 strokes/min with automatic thickness control, controlling the forging to $\frac{1}{16}$ in. That press is fed by a 4000-lb manipulator whereas the one in the laboratory at Sheffield is only a 6-cwt manipulator. Some problems increase with the size of the press and with the size of the ingot or billet owing to inertia problems.

The press at Walter Somers has been working about 15 months. A rate of 140 strokes/min may seem rather a lot. We did in fact, have some trouble a few months ago when a small screw plug got loose in the servo system and it was reduced for a time to 90 strokes/min. A year ago we should have regarded 90 strokes/min as very fast. Walter Somers, however, were complaining hard, because it made a difference between forging the billet in one heat and forging it in two; therefore we had to remove that plug as soon as possible.

Figure *A*(i) gives a record of the 500-ton press doing 75 strokes/min, with varying penetrations around ½in. Figure *A* (ii) is of the same press doing 150 blows/min. The press force is 50–100 tons. It is finishing 8 in. dia. and the average penetration is 0·15 in. The maximum error is 0·040 in. Therefore automatic thickness control is already adequate.

The forge is worked by two operators. One is at the desk, fixing the thickness control and controlling the depth of penetration, and the other is on the manipulator, moving the forging round or in and out. When I saw the press working, and when many of you saw it working at the official opening at

Walter Somers, that manipulator was being worked by a 17-year-old boy. There is some hope that we may in time be able to replace him by automatic control of the manipulator.

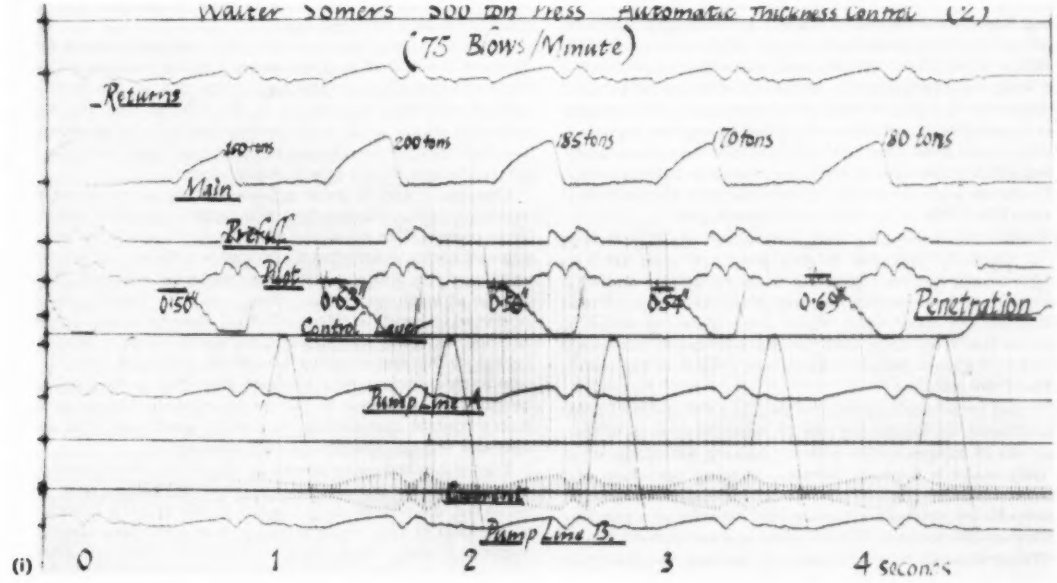
The design of manipulators must certainly be looked at. Surely we cannot continue with a manipulator which has to run around the forge, lift stuff out of the furnace, take it to the press, etc.; this is almost as crude as a muck cart on a farm.

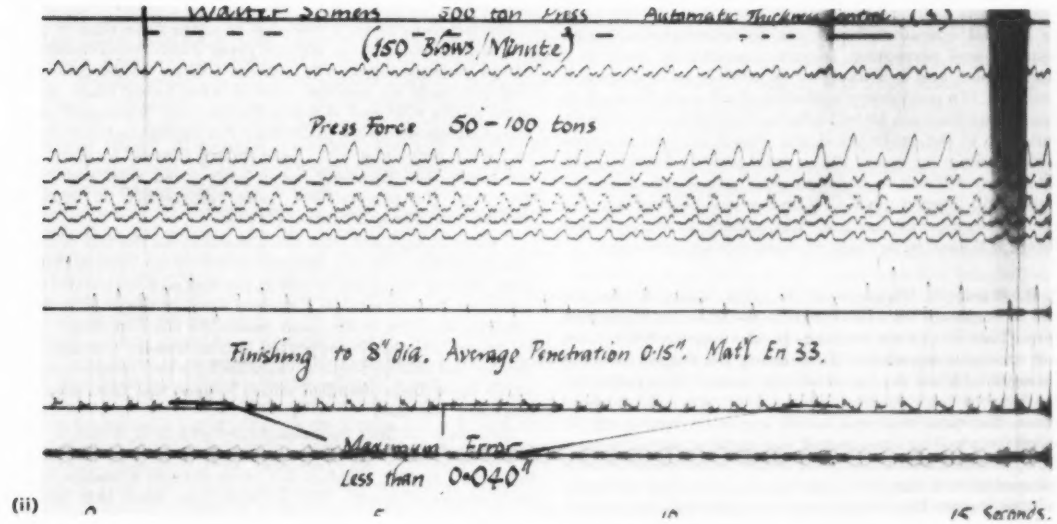
If this automatic correlation between the movements of the forging and press and the manipulator is required, I think the manipulator has to be part of the press. In other words, it is almost certain that in the end one must adopt the view that the manipulator has to be on rails and feeding into a definite position in the press, thereby being much more controllable.

The press at Sheffield was not a forging press, and never designed as such. The hydraulic controls originally supplied were manual controls which were not designed for high-speed forging and had to be modified. It is obviously much easier to achieve rapid action with a 200-ton press than with, say, a 500-ton press, and even more so than with a 1000- or 3000-ton press. There is, for instance, a difference in the weight of the moving parts to be accelerated and decelerated; and another quite important item is the decompression of the liquid in the press cylinder, which is best done in about 1/10 s.

In the small 200-ton press at Sheffield, the decompression takes place in about 0·06–0·08 s, and the rate of decompression is probably of the order of 200–300 hp. In the 500-ton press at Walter Somers, the decompression at full pressure in 1/10 s represents about 1200 hp. The decompression of a 1200-ton press, which is operating in Belgium, at full 1200 tons represents 3400 hp.

With a 3000-ton press the decompression in 1/10 s at 3000 tons is about the order of 7000 hp, which sounds a lot, but it is in 1/10 s and if it were spread over a minute would be only about 12 hp. It is not as serious as it sounds, but remains one of the limits to the speed at which a press can be operated. Returning to the diagrams of the 500-ton press that was




Walter Somers 500 ton Press Automatic Thickness Control (%)
(150 Blows/Minute)
Press Force 50–100 tons
Finishing to 8" dia. Average Penetration 0·15". Matl En 33.
Maximum Error
Less than 0·040"
(ii)
15 Seconds.

A *Record of 500-ton press operating at* (i) *75 strokes/min*, (ii) *150 strokes/min*.

working at 150 strokes/min, it was exerting only 50–100 tons, and the decompression was therefore practically nil.

There are two factors in the speed of operation. The liquid in the press cylinder must be compressed, and then decompressed before the return stroke can be started. The build-up and the decompression are affected by the pressure, and with 50–100 tons work could be carried on at 150 strokes/min. If, however, the billet had been cold, so that one was building up to 500 tons each time, the speed would have been of the order of 100 strokes/min or even less. That is one of the limitations of speed so far as presses are concerned.

Approach speed is also important. A 500-ton press may have an approach speed of 20 in./s, but the same approach on a 3000-ton press may not be so desirable. The size of the press determines the approach speed adopted.

Also the role of the manipulator must be reconsidered. The bigger the press, the bigger the billet. Whereas 6 cwt or even 4000 lb can be moved at 150 movements/min, 20 tons cannot be moved at that speed. Therefore the weight of the billet must limit speed of operation, and the manipulator will become the limiting factor on the speed of operation of the larger presses.

Until we saw this press operating at Walter Somers, I did not realize how much benefit automatic thickness control would bring by speeding up the operation of the press. I should have regarded the idea of that press working at 150 strokes/min as impossible. I certainly would not have given a guarantee for anything more than 100/min, and that very uncertainly. It is remarkable how easy operation of the press becomes when it can be set at a given size and it is certain that it will finish at that size, with little or no chance of exceeding it.

I imagine that something may very occasionally go wrong. With a manually operated forging press, an error such as making a waster and having to put back the forging and do it again, may not be reported; with an error from an automatic machine, on the other hand, there is no blame attached to anyone on the shop floor, and the machine-made mistake is reported to the management who in turn refer it to the manufacturer of the machine.

A forging press working at anything like the speeds I have indicated must do something like 10 million cycles in a year. It must do 10 million cycles without making a mistake, or at least only make a mistake during a cogging operation and before reaching the finished size. It is important to bear that in mind. Every component must be capable of operating 10 million cycles without failure. Only one component has to fail to cause damage to a large forging, damage which may be more than the equipment is worth. Therefore, every component should be scrupulously regarded for likely flaws in operation.

The manual operator should not be underrated. Human anticipation and perception is quite remarkable, even in a 17-year-old boy, and it is likely to be difficult to reproduce it in a machine. If a good driver motoring along a road sees a ball running across the road, he will take his foot off the accelerator and put it on to the brake pedal, and may even start to apply the brake because he knows that a child may be following the ball. It is very difficult to make an automatic machine anticipate what will happen next. It is well worth bearing in mind that when forging, the bar may not always go straight and somebody will have to be there to make a little correction.

Mr **J. O. Ward** (H. Wiggin and Co. Ltd): I should like the authors to comment on what seems to me to be an important problem. They make no mention in the paper of the time allowed during manipulation for hoisting the forged piece up off the anvil to allow for the set-down caused by penetration on the bottom tool. In considering our own manipulation problems, the time required to lift ingots, weighing up to 3 tons, off the anvil by a rail-bound manipulator, is of the order of 1/10 s, which is about one third of the total available time for manipulation within the press stroke. This has seriously limited the longitudinal travel and we have had to ask the manufacturers of the manipulator to increase the longitudinal speed from a proposed approximate 100 ft/min to 500 ft/min. This imposes a serious design problem of the kind to which Mr Towler referred.

The authors talk about the dimensional accuracy of the forgings and I am bound to say, from what I have seen, that their figures are correct. What I do not remember, however, is any reference to straightness. It is no good producing a forging which is precisely 6 in. square at one end when at a position 6 ft away it is twisted or offset, and one is only able to machine a $5\frac{3}{4}$-in. square bar from it. It may be that having a rail-bound manipulator produces a perfect forging, but I should like the authors' comments on this problem.

Mr **F. Mattos** (Fairey Aviation Ltd): The very first words of the excellent paper stimulate thought. Automatic equipment is certainly not very common in forging at the moment, but one wonders whether the reason given by the authors at the beginning of the paper is, in fact, the correct one, for forging will always tend to be of a non-repetitive nature. Yet in the same paper the authors have clearly put much effort into developing this remarkable automatic equipment, for which there must be a very real need.

One would imagine, therefore, that a sounder reason for the absence of automatic equipment in forging is that work of the standard described in the paper has not in the past been tackled with any earnestness in the UK, so that the forging industry, which is old and therefore tends to be conservative, has had no firm foundation on which to build into the relatively obscure realms of automatic control.

One has heard of some private efforts to introduce automatic control in conjunction with remote position control on large presses. The attempts were no doubt made by practical engineers who unfortunately lacked a sufficient knowledge of servo systems and as a result the standard achieved (I know this on at least one forging press) was such that the operator preferred to leave his selector level always on 'manual control'. Anyone watching a skilled forging press operator cannot but be struck by the amazing standards achieved without any newfangled devices such as those described in the paper, and herein lies the danger to the servo engineer. Unless an automatic control installation is a very good one, the skilled operator will not welcome it with open arms.

Figure 2 in the paper shows, in the distant background, the control desk for the BISRA experimental rolling mill, which reminded me of four days I spent in the BISRA laboratories at Sheffield in 1957 while working on an automatic strip gauge control. During those tests, virtually at the first attempt, strip gauge was controlled within $\pm 0{\cdot}0003$ in. using an electro-hydraulic servo valve in conjunction with the BISRA dyna-modulator.*

During those tests, I was able to spare some time to make the acquaintance of the 200-ton press which forms the subject of the present paper. The authors will no doubt recall that first attempts at position control were being made on the press at the time and it was suffering from 'hiccoughs' at the top of stroke which could not help but excite the curiosity of a hydraulic servo engineer. I am indeed pleased to learn from the paper that the steps taken have cured this, and indeed improved the performance quite remarkably at what the authors describe as a reasonable cost.

I have little doubt that the instability at the top of stroke was associated with the compressibility of the fluid in the pipe lines between the annulus side of the ram and the control valve which was rather remote. I wonder whether the authors would confirm that this is the main reason for the non-return valve by which 'the ram is supported in a stationary position', and that in fact this non-return valve and the two-position control valve have been installed much nearer the ram than was originally the case.

Attention to fluid compressibility effects is, I feel, a most important factor in the design of hydraulic position control systems, a matter which is not given enough attention in the early stages of design. The authors have made this point in

* BISRA unpublished report PE/R/48/57.

relation to the handling of inertia loads elsewhere in the paper, but whereas they have shown that hydraulic motors won the day on a comparison of weight, size, and torque, they have not made the most important point in a study of such a servo system, i.e. that the inertia of the moving parts of electrical machines is many times that of equivalent hydraulic machines, so that the hydraulic device is less likely to introduce troublesome critical frequencies in the operational frequency range. An interesting standard of comparison from the point of view of weight and size is that the best electrical machines cannot be operated at more than the equivalent of 200 lb/in^2, whereas hydraulic machines can operate at 4000 lb/in^2, or any other pressure one cares to stipulate, if they are correctly designed.

The problem of flexible high-pressure piping is a very real one, particularly if the flexible pipe introduces unwanted compliance between the control valve and the actuator which must do the work. Though it would not have helped in the case under discussion, there is still a need for a compact, reliable, and efficient hydraulic power pack, preferably with a servo-controlled variable delivery pump to prevent heat generation so that a large fluid reservoir need not be carried. Size and weight are very important in many cases, as indicated by the authors, so perhaps we could do worse than look towards the aircraft industry for help in this direction.

Throughout the paper, the authors have dropped repeated hints that, in some cases, they would have used better equipment had finances permitted. In the circumstances they have achieved a quite remarkable standard of control and co-ordination between the press and the manipulator, but one hopes that this principle of 'make-do' will not be extended to operational plant and that such systems will be designed right from the beginning along the lines of true servo synthesis so that all the various sub-units are accurately matched to each other to achieve the optimum system rather than one which is merely adequate. Unfortunately this basic approach to system design is more often preached than practised in heavy engineering. How often do we hear of a system costing, say, £500 being turned down in favour of an inferior one costing, say, £200 because the cheaper one will just about do. Unfortunately the compliance of the machine one is working with turns out to be somewhat different from the calculated value, so that the inferior system is now not quite good enough.

There then follow, say, 300 man-hours of experiment to win a little bit here and a little bit there to achieve the desired object, or at least to meet the specification. By any logical standard, the inferior equipment is now much more expensive and has used up senior design effort which can ill be afforded. One cannot honestly rate technical man-hours at any less than 25–30 *s.* an hour, including overheads, when one considers the standard of technician employed on servo work.

How many man-hours were spent overall in achieving the standard of autocontrol described in the paper, and from this point of view, could not the steel industry, which is by any standard a wealthy one, make more money available to BISRA for automatic control equipment as distinct from technical effort? BISRA are exercising much fine technical effort but they are, I feel, heavily hampered by lack of equipment designed specifically for the job.

For instance, the authors might not then have had to accept, as stated in the paper, 'the inevitable dead band in on-off position controls' and could have designed in an electro-hydraulic servo system for peel rotation which would have given any set angle within 1/3° without detents which limit the control to a few set positions.

High-quality multi-loop electrohydraulic servos have not been introduced in any great numbers in the steel industry so far, but these things have now passed well beyond the tinkering stage and are being widely applied to aircraft flight simulators for the training of pilots for the Britannias, Comets, etc., and many other similar ground applications. I see no reason why they should not be extended into the steel industry, where in fact they would necessarily pay large dividends. The overload case described is covered in aircraft power controls by using a pressure-loaded relief valve on each side of the jack or motor which relieves back up the pressure line when a safe pressure is exceeded.

I should like to take up a comment made by **Mr Towler** on the matter of strokes/min. I see no reason for limitation, other than the physical limitation of machines, etc. Why should any forging press not stroke at any rate one cares to specify, provided it is designed for the job? The control equipment must be there when required. For example, an aircraft now flying at Mach 2 or 3 was not conceived 15 years ago. It is possible to set a hydraulic unit oscillating at 1000 c/s. This is only a matter of designing the performance of the servo valve relative to the size of the actuator and the inertia force available.

The use of underlap valves is to be firmly avoided in most hydraulic servos. They are usually resorted to when accurate control of overlap cannot be achieved by normal techniques, but they are wasteful of energy and will change their characteristics with time because of erosion of port edges unless a very good standard of filtration is used. Manufacturers of underlap valves will protest at this, but these valves should nevertheless be classed as 'make-do's' best left alone in high-response systems. They help to linearize systems, but at the expense of stiffness with a consequent reduction in natural frequency. In any case, even though they are less amenable to calculation, non-linear systems can offer many advantages particularly where a human operator is to form part of a control system.

Servo valves are available in which overlaps can be accurately controlled within 0·0006 in. repeatedly at any required value, so that designers should not have to consider anything less accurate when designing the best thing to do the job correctly. I favour a total lap of about 0·002–0·003 in. so that a movement of 0·0015 in. from hydraulic neutral will initiate a correction at an appreciable rate. Such a valve will even position within this accuracy given a little time. Movements of 0·0002 in. can be quite easily measured and reproduced.

The results achieved by the authors are remarkably good and therefore the equipment was obviously well chosen for the task it was intended to perform. This is not to say that it would necessarily have been suitable if the sights were set somewhat higher. Double the stroke rate used for the tests is already being specified as an aim for auto-controlled presses for planishing.

The **Chairman** (Mr F. S. Somers): I should like to elaborate on a point made by Mr Mattos. In the forging industry we have to go to one manufacturer to buy a forging press and to another manufacturer to buy a manipulator; they come from two different sources. If both presses and manipulators were built by one manufacturer they could be controlled from one desk as an integral plant and many difficulties would disappear.

Mr **R. V. Rowles** (Fielding and Platt Ltd): Having seen at BISRA the equipment referred to in the paper, I congratulate all those concerned on the very great contribution they have made to the forging industry. The control and speed of penetration of the modern hydraulic press are now certainly comparable with hammer performance.

One considerable advance with the equipment under review is the very close control of thickness, which has just been stressed. When this is related to automatic control and rapidity of stroking of the press and the integration of the manipulator, it presents forgemasters with an entirely fresh concept of production.

At the risk of overlapping some of the previous papers, may I offer the following comments? In comparison with the press, the hammer depends almost wholly on the discharge of kinetic energy. Its force is therefore reduced from the moment of impact until the tup is stationary. Its manipulation requires a good degree of skill and the expenditure of physical energy, and one hears from time to time that fresh labour is not taking kindly to hammer working. However, for certain types of forgings it remains the obvious choice of equipment.

Considering the characteristics of the press: when working

direct from pumps, the maximum load can be maintained throughout the entire stroke if desired, and broadly speaking the power used is governed by the resistance of the material. With accumulator feed, regard must be paid to the usually light fall in pressure due to the expansion of the air if a gas-loaded accumulator is employed, but for large presses working at high ram speeds the accumulator is usually necessary as the installed horsepower with direct pumps becomes excessive. For instance, with presses of 3000, 5000, or 10000 tons capacity, with direct pumps a pressure of anything like that magnitude related to a reasonably high ram speed would entail installed horsepower that might not be acceptable, and an accumulator would be necessary.

Referring again to the improvised press at BISRA, one must stress the paucity of the funds that are available in the UK for this type of research. As a member of a BISRA committee one is only too well aware that adequate funds have not been available, which is surely a reflection on our national concept of fundamental research.

Bearing this in mind, one respects the work of the people who have produced the results described in the paper. It may be recalled too that this improvised press was prepared some time ago, since when the larger modern types of presses required for industry have become commercially available. The comments of Mr Somers regarding manipulators merit very serious consideration.

In conclusion, I think that what has been achieved and what has been referred to in these papers offer forgemasters a great opportunity. The modern hydraulic press, with its associated equipment, can probably make one of its greatest contributions to industry in the forging industry.

Author's reply

Commander **J. I. T. Green** (BISRA): Before saying anything about our work, I should like to repair what, I feel, is an omission in the paper: to tell you that we owe a great deal to the help of Mr Towler, especially in regard to the press, and to Mr Levetus when it came to details of the hydraulic circuit for the manipulator.

The main interest to engineers in any project of this kind is in the lessons it teaches us and is continuing to teach us. It is a pity that there was no time to build a prototype and no money to devote to development of the machine before it was wanted for forging research. It was a case of an urgent requirement arising at short notice which had to be met as quickly and as cheaply as possible.

Mr Mattos has drawn attention to the cost and effort expended on this project. We had no idea at first how much we would spend, but were thinking in terms of a few hundred pounds. This machine was designed and built within eight months of the original order to go ahead.

As regards the design, we had only our engineering instincts to guide us in its building. The rules adopted for design were lightness, strength, and stiffness in the structure, and provision for compliance or yield to protect the carriage from external forces arising from the forging process itself. We also had to bear in mind, although no precise specification was laid down originally, that reasonable accuracy would be required if good straight forgings were to be produced. We started off with the peel hanging on chains, and were faced with the problem of converting that into a mechanical system which could be controlled.

There is much talk nowadays about remote power control (rpc). It is sometimes forgotten that it is not any old thing which can be controlled by rpc, e.g. a mass suspended on a chain or a hydraulic ram actuated by a fluid susceptible to compression under pressure. Stiffness in the drive is essential for good rpc, and one problem was to provide yield under excessive external forces without departing too far from this principle.

Our early tests provided the first proof of strength in the structure. In 1957 the limit switch on the long travel failed and the resulting crash threw the machine off the rails without any apparent damage. Proof of the adequacy of yield under external forces occurred about the same time when the press got out of step with the long travel. The rubber shear pads yielded to the limit of their movement without stripping. The correct choice of stiffness has by no means been settled, and it is bound to have an effect on accuracy of longitudinal movement when high operating speeds are obtained.

In the film the manipulator looks as if it has a great deal of flexibility in it. The carriage is protected from violent shocks by the slight axial movement of the peel when the press delivers its blow.

(Then followed a 7-min film 'Automatic press forging project'.)

Mr Ward asked about the time allowed for lifting the ingot from the anvil. We allowed no time specifically for this movement. We relied on the ability of the peel to remain stationary while the carriage starts its movement. At one time we considered the addition of a hydraulic cylinder for tilting, but rejected this in the interests of simplicity.

His second question was about straightness. Up to the present, the only motion which has been given additional control of accuracy is the rotational movement. This was found necessary to avoid developing a twist in forging rectangular bars. Our solution was that described in the paper, using a detent with automatic provision for yielding when a heavy external twisting movement occurred, and subsequently restoring its position.

We now find that greater accuracy in long travel is required. A single potentiometer wire sensing position to an accuracy of 1 in 500 may not be good enough when the total travel is very large in relation to the degree of precision required. Even with continuous control, accuracy would not be any greater than it is now.

The non-return valve in the new press hydraulic circuit mentioned by Mr Mattos merely serves the purpose of locking the oil in the annulus, so that the ram remains stationary with the pumps unloaded. The condition to which Mr Mattos refers cannot happen in the new circuit owing to the presence of the prefill and exhaust valve.

In conclusion, I must emphasize that there are many possible forms which this kind of machine might have taken, but the guiding principles of design would remain the same.

WRITTEN CONTRIBUTIONS

Mr **A. F. Marshall** (BSA Group Research Centre) wrote: The achievement of BISRA in combining the operation of a hydraulic forging press and its manipulator in an automatic preselected cycle is of considerable interest.

Its application is probably limited by the generally very diverse production of the usual forge shop as ideal applications, e.g. the cogging down of special steel ingots for finishing by rolling, which occurred in US plants around 1945 at the latest, are no longer necessary.

The system might well be applied to the forging from descaled ingots of steel billets for the extrusion process, with the hydraulic forging press doing the initial cogging down, final rounding up at 50–100 tons power at 120–150 strokes/min being done by a mechanical machine, and induction heating being used to return the sawn billet to temperature before extrusion. Owing to the automatic nature of the process, little difficulty should be encountered in maintaining a protective atmosphere throughout and in providing heat reflectors where most advantageous.

I suggest that certain existing types of steam intensifier-driven presses could be utilized in automatic or semi-automatic forging. Presses of this type, made by one manufacturer at least, were rated at up to 120 strokes/min for 1000 tons capacity. The attendant turning gear was designed for the continuous rotation of its motor with the turning chain stalled intermittently by the power of the press: thus when rounding up forgings, only the press control lever need be operated. This lever was designed to give forward speed and stroke of the press proportional to speed and distance of hand lever movement. Unfortunately since the levers, links, and pins in the control system were made before needle roller bearings

became available it is doubtful how long such controls remained fully effective.

Mr **S. W. Taylor** (Wellman Smith Owen Engineering Corporation Ltd) wrote: As one of the 'engineers', referred to by Commander Green, who is responsible for the design and development of forging manipulators and other steelworks handling machines, mainly of the mobile type, I am not a forge-man, but nevertheless need to understand the requirements of forgemasters, and design forging manipulators which will withstand the arduous nature of the work in the forge, particularly if hammers are employed.

Mr Towler remarked that the automatic forge is not far away: the company I represent has on order a forging manipulator for a semi-automatic forge. The forging manipulator is not of 6-cwt capacity as is the experimental BISRA machine, but of 8-tons capacity (rated with the c.g. 5 ft 0 in. beyond the grips), and is a mobile machine presenting certain additional design problems, particularly with regard to steering. It has certain functional advantages, however, over the rail manipulator if a variety of forge work is contemplated. This machine has not only been designed, but is at the time of writing actually being assembled. It has been designed to operate: (*a*) as a fully mobile machine to work under manual control with the press, and for charging ingots into the furnace and transporting these when heated between furnace and press, (*b*) at the press by remote control on all motions, which would include long travel, steering, bar turn, grip, rocking hoist, and parallel hoist; the operation of these motions being by push buttons mounted on the press control desk, and (*c*) with the bar turn, parallel hoist, and long travel automatically linked with the stroking of the press (in this respect, the machine meets the specification of the BISRA machine, but on a very much larger scale).

It may be asked why this machine was not a rail-bound machine, as is usually envisaged for the automatic forge. The reasons are: (*a*) the mobile machine is considerably cheaper than a rail-bound machine provided with the same motions, particularly if cross traverse and/or slewing is required, (*b*) although one machine only has been ordered to date, the forge was planned to utilize two 8-ton mobile forging manipulators, one to forge while the other is fetching the next ingot (full utilization of the press can be obtained in this way; furthermore, should one machine need maintenance the plant can still function, although at a reduced rate), (*c*) the mobile machine allowed the floor area to be level and free from obstruction, and obviated the need for a pit to be provided as is usual for the rail-type manipulator, and (*d*) since the work to be forged covered a wide range, the mobile machine was considered more versatile.

This machine will be in operation, working in conjunction with a high-speed press in a new forge in Sheffield specially designed for automatic control, which should be in operation within the next 12 months, and should be of interest to forgemasters, and to BISRA in view of the excellent work they initiated. Much could be said about the detail design of a practical machine compared with the laboratory machine designed by BISRA, but would be beyond the scope of a written contribution.

However, although about 95% of forging manipulator enquiries we receive are easily satisfied by quoting for mobile machines, there is a case for the rail machine when consistent production of forgings such as axles, blocks, and bars is required. In the latter case for full utilization of the press or hammer, means of bringing the heated work to the manipulater must be provided, together with adequate furnace capacity. The question of adequate furnace capacity is of prime importance if the press and manipulators are to be fully employed for the entire shift. Large furnace capacity is not usually available, and in my experience when visiting various clients' works where a mobile forging manipulator had been put to work, although further furnace capacity had been installed to make fuller use of the forge, the manipulator had a large amount of idle time, even when transporting duties between forging were included. However, as we have a client with a forging requirement which is mainly cogging down of ingots at a high rate and under automatic control in conjunction with the press, we are designing a rail-bound machine for him. As the rail-bound machine will be tied to the press a mobile forging manipulator has been ordered to transport the ingots between the furnaces and press, and to be available to attend to special forgings outside the range of the rail machine.

The rail machine will probably be based on the mobile machine for all motions except for travelling and steering, the latter motion not being included as only a straight line travel movement is required. Furthermore, as the motion speeds of the machine must be very much higher than for the present mobile machines, the pump set complete with its oil tank and valves, and the electric motor and control gear will be placed on the floor at one side of the manipulator with flexible hoses connecting the pump set to the various motions of the machine. The machine is to be carried on free running wheels on a track provided with kick-rail to embrace both the front and the rear wheels, thereby preventing the machine jumping the tracks when operated on fast travel motion. Cylinder operation for the travel motion is envisaged, but Commander Green's remarks regarding BISRA's preference for rack and pinion drive will be borne in mind. It must, however, be mentioned that we are only able to remove the pump set, electric motor, and control gear from the machine, and provide cylinder operation for the travel motion as a travelling distance of 10 ft 0 in. is acceptable. Longer travel or larger machines may dictate putting the pump set etc. back on the machine, but not necessarily the operator; higher power would then be needed to take care of the increased weight to be travelled.

I agree with Mr Towler that although the automatic forge may have only one operator for both the press and manipulator, it is essential to treat the person as important, as skill would still be required to operate the various controls. With the complicated hydraulic operating and control gear which would be present with the automatic forge, expert and timely maintenance must be carried out, although my experience has shown that many steelworks lack enough maintenance people with the necessary hydraulic experience, which is one reason we now design our forging manipulator with the simplest possible hydraulic system.

I was very interested in Mr Mattos's remarks on hydraulic equipment and its comparison with electric equipment, as we have designed all our machines, except for the very first few, with hydraulic operation for all motions. I agree with Mr Rowles that direct pumping is more suitable than accumulators, which can be dangerous should pipes crack and spray oil towards the furnace. With regard to oil pressure, the manipulator builder is not able to operate above 2000 lb/in^2 as flexible hoses have to be employed.

Discussion on carbides and nitrides

This discussion was based on the following papers (dates of publication in the *Journal* are given in parentheses):

- **'Carbides in annealed and quenched high-speed steels'**, by *T. Malkiewicz, Z. Bojarski, and J. Foryst* (1959, **193,** Sept., 25–31).
- **'Hexagonal carbide and nitride phases and their occurrence in some alloy steels'**, by *K. W. Andrews* and *H. Hughes* (1959, **193,** Nov., 304–311).
- **'An electron diffraction method for determining the crystal lattices of small thin lamellar precipitates occurring in steels'**, by *G. R. Booker, J. Norbury,* and *A. M. Rose* (1959, **193,** Nov., 221–226).

Prof. Mgr. Ing. **T. Malkiewicz** (Institute of Ferrous Metallurgy, Gliwice, Poland) presented the first paper; Dr **K. W. Andrews** (The United Steel Companies Ltd) presented the second; and Miss **A. M. Rose** (Central Research Laboratories, Richard Thomas and Baldwins Ltd.) presented the third.

Mr **F. B. Pickering** (United Steel Companies Ltd): My contribution to the discussion will be mainly concerned with the occurrence of the M_2X phase dealt with by Drs Andrews and Hughes, but first there are some points I should like to raise in connection with the papers on carbides in high-speed steels and electron diffraction effects from lamellar precipitates.

The main point of the paper on high-speed steels is very clear: the rate of solution of M_6C is apparently much greater in the 9–4–2 steels than in the 18–4–1 type. Possibly this is due to the different positions which the two steels occupy with respect to the equilibrium system Fe–W–Cr–V–C. The authors might, with benefit, have discussed their results in the light of this system although it is undoubtedly very complex, and possibly have attempted to explain more fully this difference in the behaviour of the M_6C in the two steels.

In most cases the analyses have been carried out on residues which contain several types of carbides, and this always makes interpretation of the analysis changes very difficult. Selective removal of one or more of the phases, as used by the authors, is very useful where identifications only are required, but the doubtful quantitative nature of extraction procedures, and especially selective solution of an extracted residue, makes the resulting analytical results very dubious. Such results as are shown in Figs.7 and 8 therefore appear to me to be difficult to interpret as they comprise at least two effects, i.e. the solution of one or more of the carbides present, probably at different rates, and also the alteration in analysis of the more stable carbides by diffusion processes with both increasing time and temperature. The interaction of these factors can lead to the reversal effects shown in Fig.8.

The paper by Booker, Norbury, and Rose is an excellent example of an adaptation of selected area electron diffraction to the determination of crystal lattices of lamellar precipitates. Here we have the application of a technique similar to that used by Gard on minerals, to the examination of precipitates in extraction replicas. I should like to ask several questions on this.

It has already been mentioned that the calibration of the electron diffraction pattern alters when using the fixed angle inserts. How difficult is it to position the specimen and find the actual precipitate previously examined, and is the focus greatly affected? In the image, is it difficult to recognize a particular particle?

Secondly, would it not be possible to obtain a more or less complete pattern by dispensing with selected area and obtaining a spotty ring type of pattern from many differently oriented particles? However, the crystallographic relationship of the shape of the particle would then not be obtained.

Thirdly, the correlation of the growth of the Fe_3C in pearlite with the orthorhombic structure of the Fe_3C suggests that there may be an orientation relationship between the ferrite and the pearlitic carbide. This is supported by the obvious crystallographic shapes of extracted combs of pearlitic carbide, and certain results on the plastic deformation of pearlite which we have observed. Would the authors care to comment on this?

The authors of the paper on the M_2X type phases have said that they can form as a precipitating phase and cause hardening. I should like to present an example of M_2X occurring in this way. In a plain 12%Cr steel the first alloy carbide to form is Cr_7C_3, and this is associated with retarded softening during tempering in the range 450–500°C. Above 500°C rapid growth of the Cr_7C_3 occurs together with the formation of $M_{23}C_6$, and this results in a marked decrease in hardness. If the N content is high, however, evidence of M_2X can be observed in the electron diffraction patterns. The particle size is very small and the retarded softening is accentuated so that the steel even shows some secondary hardening. In plain 12%Cr steels much of the Cr_7C_3 forms by separate nucleation as also does the M_2X when the N content is slightly increased.

If the Cr content is increased to 17% and the steel is balanced by 4%Ni (which has little effect on the tempering reactions), Cr_7C_3 is not observed, and even at low N contents M_2X is observed to precipitate in the range 400–500°C, and overageing together with the formation of $M_{23}C_6$ occurs at high temperatures.

The M_2X phase must be Cr_2X because Cr is the only carbide-forming element present. Because there is very much more C than N, the phase is predominantly carbide, i.e. Cr_2C. Now Cr_7C_3 is slightly richer in Cr than Cr_2C, but under conditions which tend to stabilize the hexagonal M_2X structure, i.e. the presence of N or various alloying elements, it appears that a secondary hardening alloy carbide based on Cr_2C can precipitate by separate nucleation in the temperature range 400–500°C and overageing occurs at higher temperatures, with finally the formation of $M_{23}C_6$. This M_2X precipitate can cause considerable secondary hardening and its occurrence based on Cr_2C is not altogether unreasonable, because in the periodic table Cr lies vertically above Mo and W (i.e. in the same group), which both form stable hexagonal M_2C carbides while Cr forms Cr_7C_3 which is of hexagonal type. Therefore it is not surprising that the presence of N or Mo or other alloying elements can stabilize the M_2X structure, based on Cr_2C.

The addition of Mo and V to 12%Cr steels can cause M_2X to form to the complete exclusion of Cr_7C_3, and because the M_2X can vary so widely in composition, it is stable over a very wide range of temperature and therefore causes a marked increase in tempering resistance. At the lowest tempering temperature (350°C) at which we have observed M_2X it is presumably based on $(Fe.Cr)_2C$. It forms by separate nucleation and the pre-existing Fe_3C redissolves. Increasing the tempering temperature causes the M_2X to alter composition and in Mo and V steels it contains appreciably more Mo and V. This increases the secondary hardening and also causes the hardness, even after overageing, to be considerably higher because the alloy elements stabilize the M_2X to higher temperatures, which tends to keep the particle size small at any given tempering temperature. However, the maximum secondary hardening still occurs in the range 500–550°C which indicates that the M_2X, although containing some alloying elements in solution, is still based on Cr_2C. If it had been based on Mo_2C, the temperature of maximum ageing would have been 600°C.

During its formation the M_2X behaves very similarly to Mo_2C. The electron diffraction patterns show very broad lines indicating a small particle size, and the (00·2) reflections are missing until a very late stage during tempering, i.e. until a temperature of 600°C. This evidence clearly indicates that the M_2X initially forms as thin plates parallel to the basal plane of the hexagonal lattice and that the growth perpendicular to the basal plane is much slower than in other directions. This mode of formation is similar to that of Mo_2C in low-alloy steels.

Line-broadening measurements on the electron diffraction patterns can be made during the early stages of precipitation and from these the particle size and shape can be assessed.

Such measurements show that at maximum secondary hardening at 450–500°C the particle is a small plate parallel to the basal plane of the hexagonal carbide lattice, 25 Å in dia. and roughly circular in shape. Its thickness normal to the basal plane is 10–15 Å.

During tempering it grows very slowly until after tempering at 600°C it is 60 Å in diameter and 30 Å thick, and the (00·2) reflection is present in the diffraction pattern. This still small size is in agreement with the electron microstructures, and is the reason why even when overaged the hardness is very much higher than in a plain 12%Cr steel. This information therefore is supplementary to that given by the authors of the paper on M_2X phases, and places the M_2X in perspective as a member of the series of phases which act as secondary hardening precipitates in tempered steels.

Dr **H. J. Goldschmidt** (The BSA Group Research Centre, Birmingham): The paper by Drs Andrews and Hughes has brought into focus many important points concerning alloy steel constitutions. First I shall ask some factual questions on specific points in the paper, and then make a few more general observations.

Referring to Table III, what roughly is the amount of hexagonal phase in the steels mentioned, relative to that of matrix? Is it of the order of 1% or of 10% and does the phase in fact occur in such amounts as to be visible also in the steel block before extraction, either by micro- or X-ray? Or is extraction always necessary to reveal it?

An analysis of the residues as distinct from that of the bulk steels in Table III would have been interesting, at least in one or two cases where the hexagonal phase predominates.

These hexagonal phases are very important because of their structure relation to the cubic carbides, and to the austenitic or ferritic matrix phases. In many ways they can be regarded as a hybrid condition between face-centred cubic austenite and the true MX carbide; in this sense, the hexagonal symmetry is adopted only as a transitional compromise in an endeavour to maintain close-packing.

The possibility of stacking fault formation should be considered. This would give, within a single parent crystal, fluctuating lattice-regions of cubic and hexagonal type layer-repeat, depending on local segregation of interstitial atoms or of vacancies. It is arguable where the boundary lies here between 'matrix' and 'carbide', and the term 'hybrid carbide' might be appropriate; one might even call it a 'matride', which might be defined as an interstitial phase, which can form through coherent precipitation from the matrix with little, if any, composition change, but with such change in lattice symmetry as would favour potential further solid solution of interstitial atoms, towards a carbide or nitride. This is not necessarily restricted to hexagonal phases. We have in fact observed a carbon-supersaturated austenite, which is already very much akin to a carbide; though cubic, it differs in composition only little from the hexagonal M_2X.

What is the authors' evidence for M_3X being the end-composition in defect lattice formation? Could it not be still metal-richer, so as to overlap quite widely with known austenite compositions? Alternatively the significance of M_3X might be that it marks the approximate saturation limit of the austenite for (C+N) in these steels.

This raises another point. Is it not conceivable that in the actual bulk steel the hexagonal carbide might sometimes not exist, but could be created by the electrolytic extraction process? Conceivably the difference in free energy is so little that extraction would have a critical influence, and a slight preferential leaching out of one or the other element from the matrix leaves interstitially loaded regions, which could switch into the hexagonal form. This is a possibility that might be investigated, and gives some point to my earlier question.

The transition cubic–hexagonal close-packed is not confined to interstitial alloys but also occurs in pure metals. The case of cobalt comes immediately to mind, and with this element being next on the right to iron in the periodic table in Group VIII, one might well say that the addition of interstitial carbon or nitrogen to iron-base austenite renders this in fact more 'cobalt-like', electronically and structurally, by supplying the d-shell of the iron atom with the extra electron needed to produce the cobalt configuration. In this sense, the cubic austenite and the hexagonal M_2X phases are in fact in a relation quite analogous to cubic and hexagonal cobalt. This link with cobalt makes stacking fault formation in the austenitic matrix and in the derived carbides all the more likely.

It is also extremely interesting here to note the recent discovery by Sandor and Wooster[1] that vanadium single crystals containing very little nitrogen in solution show an effect upon Laue spots which can be explained by the formation of a hexagonal phase. Although in this case it is formed from the body-centred cube, the phenomenon becomes very significant if we also consider ferrite, as we must. There is no time to enlarge on this now, but I should just like to make the point that, first, the stabilization of the hexagonal structure by interstitials is a basic occurrence of much wider scope. Secondly, since a hexagonal vanadium naturally brings one to titanium, one might ask rather provocatively whether even in pure titanium the bcc/hexagonal allotropy perhaps occurs only through the presence of minute traces of interstitials and whether in their absolute absence the hexagonal α-Ti form might not be stable?

Dr **K. W. Andrews** (United Steel Companies Ltd), in reply: With regard to the amounts of phases in Table III, there is an acknowledged weakness in our experimental methods. A fully representative residue and complete analysis are to some extent incompatible. If one requires an adequate amount of residue for analysis then one is likely to lose the finer particles. We prefer at present to get all the particles we can and to fill in the gap with correlations of the kind we have used here. We have tried to develop a method for the analysis of such residues and we shall be talking about that this afternoon.

I had considered Dr Goldschmidt's notion that hexagonal phases are hybrid. This has not been properly appreciated, but in a German paper on the precipitation of ε-carbide I have noticed that the hexagonal lattice could be regarded as being about half-way between the two cubic phases in the mode of transformation suggested. There are different ways of looking at the same transformations. He also mentioned the possibility of lack of definition between carbide particle and matrix, and we are bearing this in mind. There are times when one cannot certainly differentiate the matrix from the carbide, and this is a valid point. When there are very fine particles coherent with the matrix it is problematical where one leaves off and the other begins.

Dr Goldschmidt asked about the evidence of the M_3X end composition. If one looks at the diagrams, M_2X often seems to be one limit and some diagrams show the phase stable as far as M_3X. More often, however, they are stable at M_2X but do not extend down to M_3X.

He also asked whether M_2X could be created by the extraction process. It would be very interesting if it could, and I agree it would be worth investigating. In our own work, we are convinced that these carbides are there in the solid steel. Mo_2C and other M_2X phases have often been seen under the optical or electron microscope, for example.

It was suggested that one was rendering iron more cobalt-like by adding C or N. I have heard it said that iron has narrowly escaped being hexagonal. This is a view supported by theoretical calculation. There may well be some truth in these suggestions but I think that by adding interstitial atoms the conditions are altered considerably. Placing atoms in interstitial instead of substitutional positions alters the electronic distribution and adds lattice strain energy, thus altering considerably the circumstances which might favour one phase or another. We briefly examined some of these theoretical interpretations but we felt that we could say nothing new beyond what has already been said in other papers on the MX phases. We are not certain about some of these interpretations and how far we should take them.

TABLE 1 Carbides in annealed and hardened specimens

Steel	Chemical analysis, % C	W	Mo	Cr	V	Co	Carbides in the annealed steel Wt-% of total steel	Carbide type and amount	Carbides in the steel as-hardened Wt-% of total steel	Carbide type and amount
18–4–1	0·71	17·8	0·46	3·95	1·10	...	29·0	M_6C(80%) + ($M_{23}C_6$ + M_7C_3)(15%) + MC(< 4%)	Hardened 1 290°C 17·0	M_6C(100%)
6–5–4–2	0·75	6·1	4·20	3·75	1·40	...	20·6	M_6C(80%) + MC(10%) + $M_{23}C_6$(10%)	Hardened 1 180°C 14·8 Hardened 1 220°C 13·3 Hardened 1 240°C 13·0	M_6C(96%) + MC(4%)
14–4–4	1·27	13·3	0·34	4·65	3·70	...	26·8	M_6C(60%) + MC(25%) + $M_{23}C_6$(15%)	Hardened 1 220°C 20·0 Hardened 1 250°C 19·1 Hardened 1 270°C 19·1	MC(50%) + M_6C(50%)
12–4–5–5(Co)	1·46	13·0	0·32	4·75	5·05	5·8	25·9	MC(48%) + M_6C(45%) + $M_{23}C_6$(10%)	Hardened 1 250°C 16·5	MC(80%) + M_6C(20%)

I was interested in the work of Sandor and Wooster and was gratified that they found a hexagonal V_2N precipitate instead of the cubic ones in the cubic lattice. Dr Goldschmidt's final point about titanium is also most intriguing. In the future we shall take even more note of the fact that the hexagonal lattices are so close to the cubic that we ought to find them more often than we do.

Dr **S. G. Cope** (Consett Iron Co.): Some years ago I investigated the carbides in four different high-speed steels, 18–4–1, 6–5–4–2, and two 4% vanadium steels, using, among other techniques, electrolytic extraction, X-ray examination, chemical analysis, and hardness determinations similar to those described in the paper by Malkiewicz *et al.* I also made particular use of differential etching. The methods used were in general similar to those of Blickwede and Cohen.[2] I concentrated mainly on tempered structures using the annealed and quenched structures only as a background, because Kayser and Cohen[3] had just reported comprehensive investigations on carbides in many high-speed steels in the annealed state and after various times and temperatures of hardening, and seemed to have covered this field well.

My aim was to study the carbides precipitating during tempering by means of X-ray examination of extracted carbides and by differential etching of treated steels. The metallographic approach was unsuitable for the detection of early stage precipitation, but in the late stages of tempering, when the annealed structure is being approached, the appearance of new carbides and their pattern of growth can be detected metallographically. This has some interest relative to the annealed structure itself and accordingly some of my findings in the late stages of tempering as well as in the annealed and as-hardened state are now presented.

Table 1 shows the analyses of the high-speed steels used and the main findings concerning carbides in the annealed and quenched conditions. The percentages of total carbide shown represent the mixed carbides recovered by electrolytic extraction. The estimated percentages of the different carbides are based on X-ray indications.

Of the minor phases, the presence of MC, the vanadium-rich carbide, is quite definite in the 6–5–4–2 steel. In 18–4–1 steel it appeared only very faintly, close to the limit of X-ray detection. The chromium-rich minor phase was present with certainty in all four steels in the annealed state and at first was assumed to be $M_{23}C_6$. At the time, I was not aware of the procedure for separation of carbides put forward by Popova, and the identification of the $M_{23}C_6$ was based unsoundly only on one or two faint lines. Subsequently, going back over a stock of extracted carbides and using the Popova technique, I have been able to confirm that in my annealed 18–4–1 steel, $M_{23}C_6$ and M_7C_3 are both definitely present in small quantities. In the other three steels only $M_{23}C_6$ appears.

The Popova procedure worked very well in the removal of M_6C and, in the case of 18–4–1 and 6–5–4–2 steel, where MC appears only as a minor phase, this latter carbide also is removed; but in the case of the two high-vanadium steels where MC appears as a major phase, very little solution of this carbide takes place although M_6C is dissolved completely. The presence of a large quantity of MC in these cases makes positive identification of the minor chromium-rich phases rather difficult in spite of the removal of M_6C.

Since in most cases only one hardening temperature was used, I have no coverage of the effect of variation of austenitizing conditions such as have been presented by Malkiewicz, Bojarski, and Foryst. However, in preliminary work, three different hardening temperatures were used for two of the steels. The effects of these are shown in Table 1. The range of temperatures employed was small and the effect is shown only by a difference in the total quantity of carbide remaining undissolved. None of the chromium-rich carbides appears in any of the hardened specimens so it must be assumed that all had dissolved at a lower temperature than any employed for hardening.

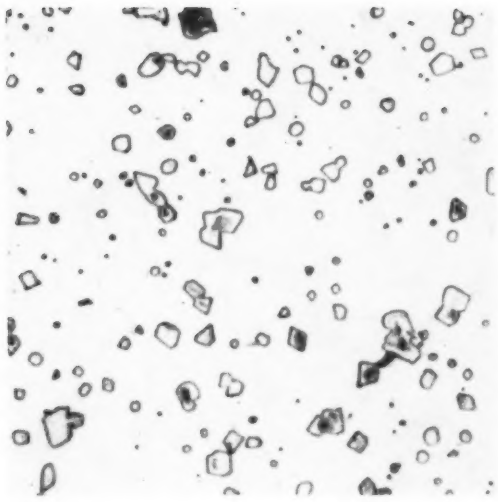
Table 2 shows examples of the X-ray analysis of carbides extracted from steels tempered at high temperatures, approaching the annealing temperature. In both cases shown it can be seen that the first carbide to precipitate is one of Dr Andrews' hexagonal carbide phases referred to in Table 2, rather loosely, as W_2C. This is a transient phase which appears always only as a diffuse X-ray pattern and disappears on further tempering.

The order of appearance of subsequent carbides is difficult

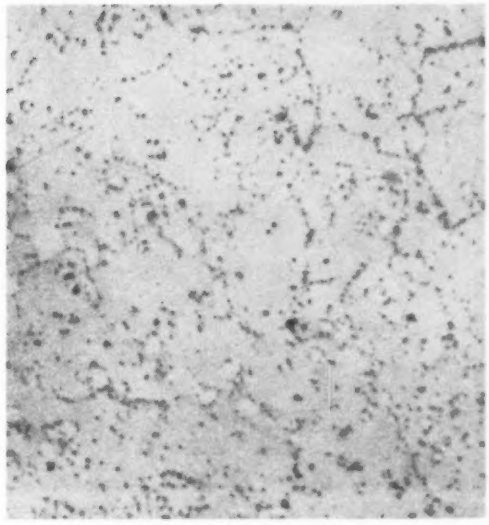
TABLE 2 Carbides in tempered structures (X-ray results)

Condition		18–4–1 steel	6–5–4–2 steel
As-hardened		M_6C	M_6C + MC(w)
Tempered 700°C	15 min	W_2C(D)*(ss)	W_2C(D)(m)
	30 min	W_2C(D)(ss)	W_2C(D)(ss)
	1 h	W_2C(D)(ss) + M_6C(B)†	W_2C(D)(ss)
	2 h	W_2C(D)(m) + M_6C(B)	W_2C(D)(ss)
	5 h	W_2C(D)(w) + M_6C(B) + $M_{23}C_6$ and/or M_7C_3(vw)	W_2C(D)(w) + M_6C(B) + MC(B) + $M_{23}C_6$(D)(w)
	50 h	M_6C + $M_{23}C_6$(vw) + M_7C_3(vw)	M_6C + MC(w) + $M_{23}C_6$(w)
750°C	1 h	W_2C(D)(w) + M_6C	W_2C(D)(w) + M_6C + MC(w) + $M_{23}C_6$
800°C	1 h	M_6C + $M_{23}C_6$(vw) + M_7C_3(vw)	M_6C + MC(m) + $M_{23}C_6$(m)
	10 h	M_6C + $M_{23}C_6$(w) + M_7C_3(w)	M_6C + MC(m) + $M_{23}C_6$(m)
Annealed		M_6C + $M_{23}C_6$ + M_7C_3(w) + MC(w)	M_6C + MC(m) + $M_{23}C_6$(m)

Note: all tempered structures contain the undissolved carbides of the hardened structure. The carbides listed are those precipitated in tempering.
*(D) diffuse line; †(B) broadening of an existing line.



B *Hardened 18–4–1 steel etched consecutively in* $KMnO_4$ *and* Cr_2O_3



D *18–4–1 steel, tempered 5 h at 700°C, etched in* Cr_2O_3

to determine with certainty, but the order shown in Table 2 for 18–4–1 and 6–5–4–2 steels is fairly well defined. There is no sign in any of these steels of an intermediate appearance of any carbide other than W_2C, which does not persist in the annealed structure.

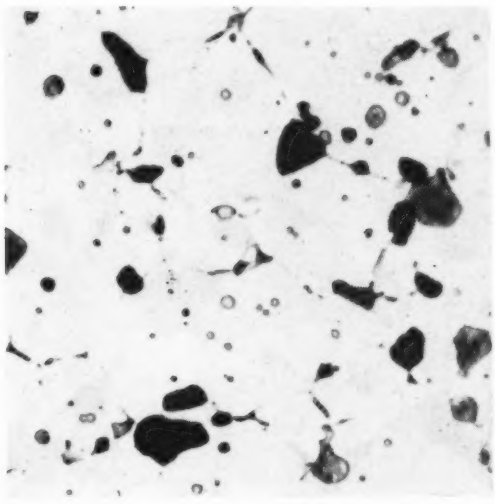
Blickwede and Cohen[2] made considerable use of selective etching. I employed their etchants and also tried numerous other etchants in an attempt to build up a metallographic picture of carbide precipitation and growth in the late stages of tempering.

Figure *B* shows hardened 18–4–1 steel etched consecutively with potassium permanganate and chromic acid. The carbide outlined is M_6C etched by the permanganate; no other carbide is present.

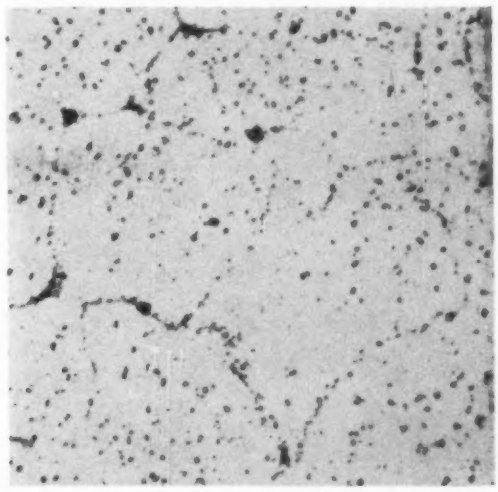
Figure *C* shows 14–4–4 steel etched in the same way. Here the M_6C is seen etched by permanganate with an interference film while the MC carbide is clearly differentiated by black etching with chromic acid.

Metallographic indications in all quenched steels match well with X-ray examination of extracted carbides. This is to be expected since all carbides remaining undissolved are relatively large.

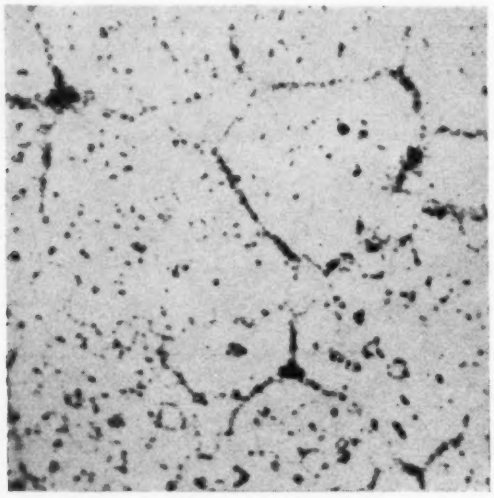
In the case of the tempered steels a correlation between microscopic examination and X-ray examination carbide was not possible. The first appearance of a new phase precipitating during tempering in all four steels was concentrated at grain boundaries and around existing coarse carbides. It showed the etching characteristics of MC, i.e. it was shown up by chromic acid but not by permanganate. I do not offer any explanation of this because there was no obvious appearance of MC in the extracted carbides. In papers about secondary hardening it has been claimed that M_2C is a main cause of precipitation hardening in high-speed steels. All-over fine precipitation of MC has also been suggested as a means of bringing about secondary hardening, and electron microscope studies show that in certain cases this all-over precipitation appearing at an early stage changes later to a growth of particles and a concentration around the grain boundaries. Figures *D–F* show examples of this. Figure *G* shows how the grain boundary precipitate still shows up against the all-over precipitate of M_6C which has commenced at a later stage of tempering.



C *14–4–4 steel etched consecutively in* $KMnO_4$ *and* Cr_2O_3



E *18–4–1 steel, tempered 50 h at 700°C, etched in* Cr_2O_3



F *6–5–4–2 steel, tempered 50 h at 700°C, etched in* Cr_2O_3

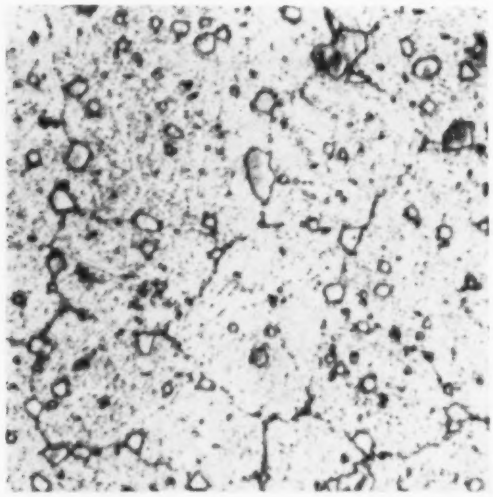
The appearance of M_6C at a later stage of tempering than the grain boundary precipitate (MC?) is shown in Figs. *H–J*.

Figure *H* reveals no sign of newly precipitated M_6C. In Fig. *J* the general background precipitation of M_6C is now obvious. This sequence is continued in Figs. *K–M*.

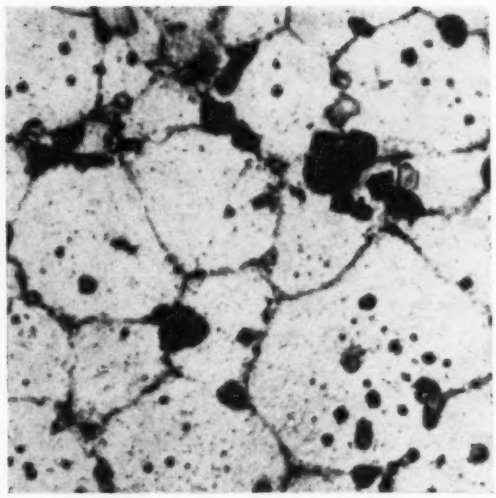
Figures *N* and *O* are shown as examples of the selectiveness of the etchants. The blank areas represent unetched MC particles. Comparison of these microstructures with similar ones additionally etched in chromic acid indicates that the background precipitate is entirely or almost entirely of M_6C.

A disappointing feature of the selective etching was the failure to find direct indications of the distribution of $M_{23}C_6$ (or M_7C_3) in spite of trials with several different etchants, such as oxalic acid, which give good results with carbides in stainless and similar steels. In annealed structures the M_6C was estimated by differences between full carbide etching with nital and combined M_6C and MC etching. In fully tempered specimens this is not possible, because of background darkening.

I should be interested to hear whether the authors have any experience to offer in the metallographic distribution of different carbides in the annealed structure.



G *18–4–1 steel, tempered 50 h at 700°C, etched consecutively in* Cr_2O_3 *and* $KMnO_4$



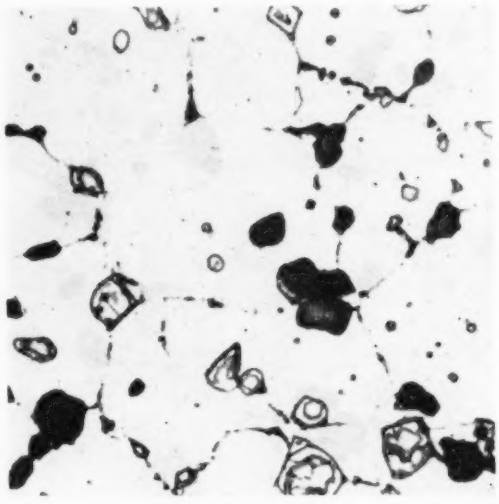
H *14–4–4 steel, tempered 1 h at 700°C, etched as G*

I hope that it is not thought that my reference to tempered structures is irrelevant to the present discussion. I feel that they are of interest because their continuation defines the annealed structure. The distribution and pattern of the different carbides in the annealed structure is of some importance.

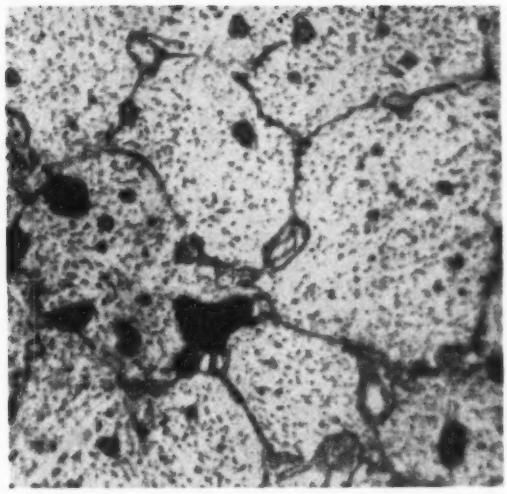
In other work which I completed, directly comparable with that of the present paper, e.g. chemical analysis of extract and matrix, my results were generally in agreement with those of Messrs Malkiewicz, Bojarski, and Foryst.

Mr **G. Hoyle** (BISRA): We have done a limited amount of work on high-speed steel carbides and I would like to raise two points in connection with the paper on this subject. All my remarks refer to 18–4–1 steel.

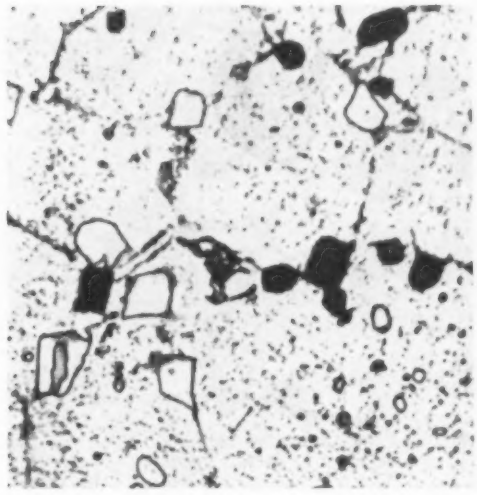
First, we confirm the authors' views on contamination of the residue from electrolysis by metallic particles at high current density. We were able to produce a residue that was completely non-magnetic at current densities of 0·02 A/cm². Comparison of the weight of residue from high and low current density extractions suggested that the metallic contamination accounted for about 1–2% of the residue.



I *14–4–4 steel, tempered 5 h at 700°C, etched as G*



J *14–4–4 steel, tempered 50 h at 700°C, etched as G*



L *14–4–4 steel, tempered 10 h at 800°C, etched as K*

In our experiments we were intending to use Blickwede and Cohen's magnetic separation technique to purify the residues, but we found that more residue was extracted magnetically than would be expected, and a check showed that this magnetic portion which had been discarded by Blickwede, was substantially η carbide.

Attempts to purify residues showed that hydrochloric acid certainly dissolves a substantial portion of the residue. Four successive treatments gave an appreciable loss of weight, and we discarded the method as being unsuitable.

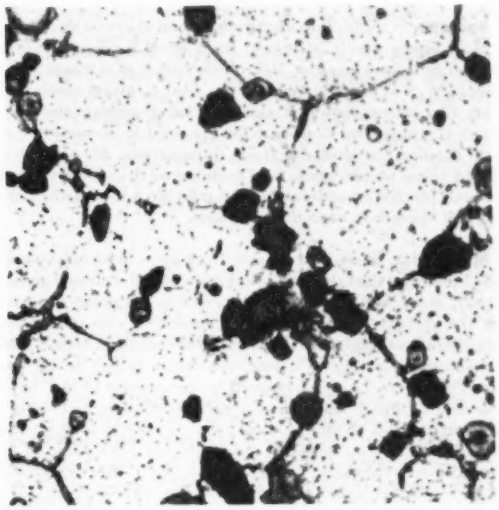
Our results for the carbide residue were rather peculiar and suggested that the weight of undissolved carbide fell only from 26% to 25% as the austenitizing temperature was raised from 1200° to 1320°C, a result so much in contradiction with recognized theories that, although by X-rays we could not find anything but η-carbide in the residues, we suspected some other contamination. Have the authors any comments to offer on this?

My second point concerns the lattice parameter of η-carbide. The paper discusses the lack of agreement concerning composition changes in η-carbide with austenitizing temperature and suggests that any such changes may be connected with the formation of polygonal carbides. Our results gave a constant value of 11·032 kX units for η-carbide in 18–4–1 steel annealed or quenched from 1200° or 1260°C. Samples of the same steel quenched from 1320°C gave 11·04 kX units.

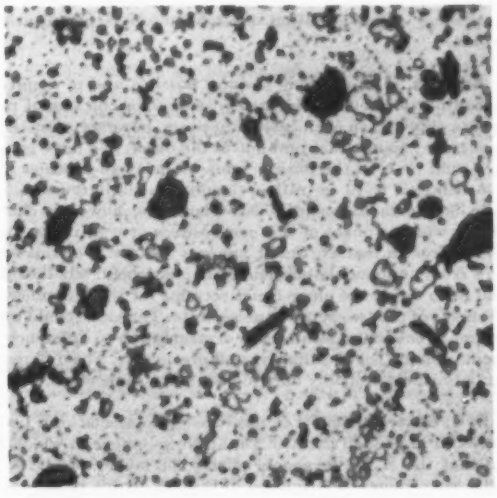
Our austenitizing time was only 45 s and this may account for the fact that we did not find any change at 1260°C; with the authors' longer treatments polygonization of carbides was reported at 1250°C.

On the other hand, the idea of a constant lattice parameter at lower temperatures, below 1310°C, and a change above this temperature, agree with some of our observations. It is universally accepted that the attainment of equilibrium in high-speed steel is a very lengthy operation. Figure 3 in the paper suggests that equilibrium might be attained in 30 min at 1300°C. This is for a fully worked structure, and an as-cast structure, with eutectic networks, would not be expected to behave in the same way.

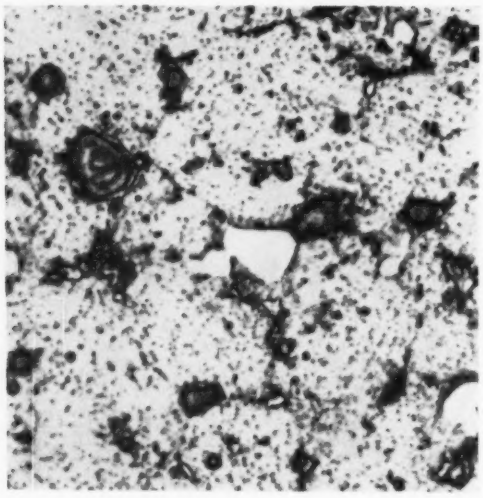
In the range 1310–1330°C we find that a 30-min treatment gives virtually the same microstructure with polygonal car-



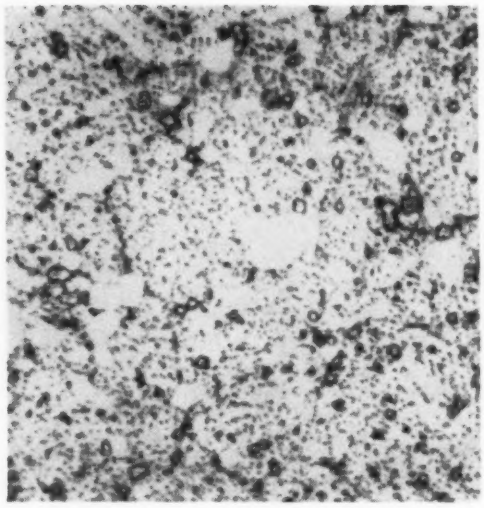
K *14–4–4 steel, tempered 1 h at 800°C, etched consecutively in* Cr_2O_3 *and* $KMnO_4$



M *14–4–4 steel, fully annealed structure, etched as K*



N *14–4–4 steel, tempered 50 h at 700°C, etched in* $KMnO_4$



O *12–4–4–5 steel, tempered 50 h at 700°C, etched as N*

bides and no eutectic, irrespective of the initial condition of the specimen.[4] We believe the change in structure is due to the peritectic reaction, and assume that the final polygonal carbide structure represents true equilibrium. Can it be, then, that two mechanisms for solution of carbide are operative: one, at low temperatures, not accompanied by a change in parameter or composition, and one, at 1310–1330°C, with a change?

Mr **D. T. Llewellyn** (United Steel Companies Ltd): Dr Andrews has shown that in the M_2X type of crystal, the nitrides have high axial ratios mostly between 1·60 and 1·62, while carbides have axial ratios between 1·56 and 1·60. In ferritic stainless steels we have observed the precipitation of Cr_2N, after reheating at 400°C for 1 h, in the form of long needles (0·001 in.). This precipitate causes a slight decrease in hardness and it is considered that the loss in alloying elements in solid solution caused by precipitation offsets any hardening effect that the precipitate might produce. Mr Pickering has shown that the M_2X precipitate formed after similar heat-treatment in 12%Cr steels, which is probably based on carbide, produces a marked secondary hardening effect and is very fine, being of the order of 200–500 Å. Could it be that the difference in axial ratios of carbides *v.* nitrides is responsible for the difference in the morphology of precipitation of Cr_2N and M_2X based on the carbide?

Dr **Andrews,** in reply: This is quite possible and it would be interesting to calculate the lattice strains for the size of precipitate observed in each case. It appears, however, that there is coherency in one case and not in the other. We have generally found Cr_2N in the form of somewhat larger particles and have believed it to be rather 'inactive' metallurgically, i.e. more like an inclusion (of TiN). There are, however, conditions in which a phase of this composition or near it, can precipitate from a matrix, but it appears that even in this case there is no appreciable hardening. It is doubtful whether the axial ratio difference alone could account for the difference in the effects of the two types of precipitate. Thus, one could take account of coherence strains and of strains due to difference in specific volume between matrix and precipitate. The difference in axial ratio is probably only incidental to these effects.

Dr **R. G. Baker** (British Welding Research Association): I was interested in a small point in Dr Andrews' paper in which he said that the M_2X (hexagonal phase) can sometimes be seen in the early stages of formation of the MX (cubic phase), and I should like to suggest a possible reason why that might occur. Taking the case of vanadium carbide in ferrite one finds that the precipitate occurs as plates. The orientation relationship appears to be $\{100\}_{VC} // \{100\}_{\alpha\ iron}$. $\langle 100\rangle_{VC} // \langle 110\rangle_{\alpha\ iron}$. The misfit is about 2% in $[100]_{VC}$ and $[010]_{VC}$ directions, but in the $[001]_{VC}$ direction (perpendicular to the plane of the plate) the misfit is 30%. As the plate continues to grow it would be expected to break away from the lattice when or before the total misfit reaches 100%, i.e. when the plate is two or three layers of atoms thick.

Some of the considerable strain energy which is produced in both the plate and the surrounding matrix might be relieved if the face-centred cubic VC sheared on every third (111) plane to produce a hexagonal structure. I am not suggesting that this is in fact what would happen. One would expect rather that the precipitate might grow as hexagonal structure in the first instance if the strain energy thus relieved were greater than the total stacking fault energy introduced. (A hexagonal structure may be regarded as a face-centred cubic structure with a stacking fault on every third plane.) When the precipitate became non-coherent the strain energy resulting from misfit in the [001] would be relieved at first and the precipitate might then revert to the face-centred cubic structure.

Dr **Andrews,** in reply: I do not think we can claim to have found an M_2X phase which is formed from an MX phase. Our impression is that when the MX phase forms it remains stable whereas M_2X is often transient. We need to know much more about the orientation relationship during precipitation and I think the paper by Booker, Norbury, and Rose presents one useful method for studying these orientations. We must bear in mind these points made by Dr Baker as they seem to me to be worth further consideration and provide a mechanism that may well operate in certain conditions.

Mr **Malkiewicz,** in reply: I cannot fully agree with the opinion of Mr Pickering that the rate of solution of M_6C is much greater in 9–4–2 steel than in 18–4–1. Figure 11 in our paper indicates that the differences in tungsten content of the matrix are negligible in both steels. Realizing that the solution of M_6C must be accounted for, the solution rate seems rather to be very close. This is not the case in regard to the amount of M_6C in the steels in question, both in the annealed and quenched state.

The separation of different carbides in the electrolytic residue is difficult. We succeeded only with the separation of the chromium-rich carbides, i.e. $M_{23}C_6$ and M_7C_3. The determination of the composition of carbides in the residue from 9–4–2 steel M_6C+MC is impossible, but we have analysed M_6C in 18–4–1 steel, austenitized at high temperatures, when only M_6C is present, and have found the following figures:

Austenitizing temp., °C/time, min	W	Cr	V	Fe	C
1200/20	65·07	3·67	2·98	25·01	2·28
1300/60	66·67	3·50	2·70	24·98	2·27
1300/120	66·20	3·49	2·85	24·52	2·18

The carbon content is considered to be too high. The residue from electrolytic extraction is contaminated with colloidal carbon from the dissolution of the martensite, and this carbon is very difficult to separate.

The M_6C carbides in the 18–4–1 steel, at high temperatures and with longer soaking times, are growing into few, but large, angular particles. Measuring their lattice parameters we have found that they are increasing during the first 10–20 min of soaking but afterwards remain practically constant. The quoted differences in the chemical composition may be partly responsible for this change in parameter. We suppose that the carbides have a cored structure, because the finer particles are dissolved during coagulation and precipitated on the growing particles from a matrix saturated to about 10%W, which is much more than at lower temperatures. Thus the outer layers of the growing carbide particles reflecting the X-rays can have a higher tungsten content than the average given by the chemical analysis.

The differences in the chromium-rich carbides in the annealed steel, relative to the figures given by Mr Cope in Table 1, may be due to the differences in annealing. If the annealing treatment is carried out at 730° or 850°C different carbides may precipitate. We have not used differential etching and have no experience with this technique.

Mr Hoyle has spoken about the polygonization of carbides. I suggest the following explanation. At high temperatures there is a strong tendency to coagulation of M_6C because the rate of diffusion of tungsten increases appreciably. The interface energy plays an important role. At temperatures in question the matrix is near the melting point, therefore the surface energy of the carbide crystals may be the decisive factor for the angular shape of the precipitate. I think this is the approach to equilibrium, quoted by Mr Hoyle, without passing through the peritectic reaction, which can take place at higher temperatures only. The angular shape of carbides results from conditions of high-temperature coagulation.

Referring to Mr Hoyle's question about the very small differences in the amount of undissolved carbide at 1200° and 1320°C, I suggest that the used austenitizing time, only 45 s, was too short for dissolution of carbides. We have found for samples of 18–4–1, 10 mm in dia., soaked at 1250°C, differences of up to 2% in the amount of undissolved carbides when the austenitizing time was increased from 1 to 10 min.

Mr **G. R. Booker,** Dr **J. Norbury,** and Miss **A. M. Rose** wrote: Mr Pickering enquires about finding a particular particle previously examined when an extraction replica is transferred from one fixed-angle insert to another. This is facilitated by beginning each examination at a low magnification of about ×75 obtained by withdrawing the contrast aperture, switching off the objective lens and focusing with the first projector lens. The portion of the particular grid opening containing the particle of interest can then be readily moved to the centre of the imaging screen. The magnification is then re-set to about ×1000, the image focused and the area of interest again centred. The magnification is then increased to, say, ×10000, the image focused and the particle of interest located and centred. The particle is generally at a slightly different distance from the objective lens than when observed in the previous insert. Consequently, the setting of the objective lens current in the two instances is slightly different for the two focused images at the final ×10000 magnification, and the electron diffraction calibration constants are slightly different. The former has little significance, but the latter should be taken into account, possibly by the method described in the paper, when accurate results are required.

TABLE 3 Spacing of atoms in the cementite and austenite crystal lattices

Cementite direction	Spacing AU	Austenite direction	Spacing AU	% Misfit
[100]	4·51	$[\bar{1}\bar{1}2]$	4·41	$2\frac{1}{2}$
[010]	5·08	$[1\bar{1}0]$	$\frac{1}{2}\times 5{\cdot}08$	0
[001]	6·72	[111]	6·23	9

The possibility suggested by Mr Pickering that additional useful information might be obtained by dispensing with selected-area electron diffraction and obtaining a spotty ring type of pattern from many differently oriented particles has been confirmed in practice. This mode of examination has been performed either by slowly scanning the extraction replica over small distances during the electron diffraction exposure, as described in the paper, or by using selected-area diffraction with a relatively large limiting aperture. Both these methods have enabled strong diffraction rings to be recorded from particles in tilted extraction replicas when the rings were absent or weak in patterns similarly obtained from untilted replicas.

Mr Pickering's suggestion that pearlitic carbides may grow in the matrix in a Widmanstätten manner is in accord with our views. Attempts to find a simple relationship between the orthorhombic Fe_3C lattice and the ferrite lattice have been unsuccessful, but such a relationship has been found between the Fe_3C lattice and the austenite lattice (*see* Table 3). The small atom misfits occurring in the $[100]_C$ and $[010]_C$ directions would favour the formation of plate-like particles with the $[001]_C$ direction perpendicular to the plane of the plate, in agreement with the experimental observations. The significance of these observations has yet to be evaluated.

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2. D. J. Blickwede and M. Cohen: *Trans. AIME*, 1949, **185,** 578–584.
3. F. Kayser and M. Cohen: *Met. Prog.*, 1952, **71,** (6), 79–85.
4. G. Hoyle and E. Ineson: *JISI*, 1959, **193,** 254–269.

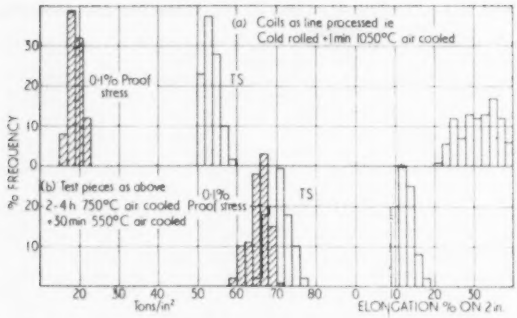
Discussion on controlled transformation

This discussion was based on the following paper (date of publication in the *Journal* is given in parenthesis):

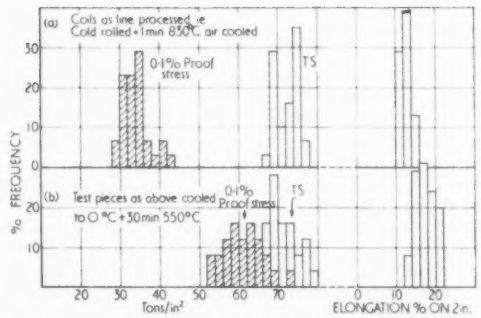
'Controlled-transformation stainless steels', by *K. J. Irvine, D. T. Llewellyn,* and *F. B. Pickering* (1959, **192,** July, 218–238).

Dr **K. J. Irvine,** who presented the paper, said it had a bearing on the discussion which had taken place in the discussion on carbides and nitrides on how research work should be tackled. The authors had had to leave many interesting facets of the work for later explanation but felt that they had the broad picture which enabled them to carry on with the development of the steels. Had they dealt with the matter the other way round they would still be doing the preliminary work and no steel would have resulted from it.

Mr **J. I. Morley** (Firth–Vickers Stainless Steels Ltd): Not many years ago there were metallurgists who confessed to being puzzled



P *Results of 104 transverse tensile tests (15 casts) representing 21 coils 24–36 in. wide, austenitic strip 0·020–0·080 in. thick*



Q *Results of 31 transverse tensile tests (8 casts) representing 12 coils 24–36 in. wide, martensitic strip 0·035–0·080 in. thick*

by variations in yield strength of the simple 18/8 titanium steels from one cast to another. Some were reluctant to attribute these differences to slight differences in composition, for example the difference between 18%Cr and 8%Ni on the one hand and 17½%Cr and 7½%Ni on the other. At this time a great deal of work was done on the effects of adding other elements to 18/8 steel, but steels such as those which have been so well studied and described by the present authors were regarded as belonging to a sort of no-man's-land of instability.

We must, I think, acknowledge the pioneer work done in the USA in opening up this field. A US paper on the subject was published in 1931 and before 1950 we were beginning to read in US trade journals and patent literature of basic compositions such as 17%Cr–4%Ni, with a martensitic type of structure, and 17%Cr–7%Ni, with an unstable austenitic structure, and with the addition of precipitation-hardening elements such as copper and aluminium. As sources of technical information these references left much to be desired. Progress with the development and use of these so-called precipitation-hardening steels has gone on uninterruptedly in the USA ever since 1950. The authors have entered this field of investigation somewhat late but, so far as I know, there is not a single paper in US literature dealing with the metallography and the specific effects of composition so fully as is done in this paper.

I commend the use of the term 'controlled-transformation steels'. These steels have been called by other names, such as 'quasi-martensitic steels', 'semi-austenitic steels', and so on, but such names do not recognize their most important single characteristic, that of having the transformation temperature just a little below room temperature.

Some understanding of the metallurgical characteristics of these materials by the user is essential. I wish the authors could have gone a little further in this direction and told us of the harmful effects of carburizing and nitriding during heat-treatment cycles, which can result in incomplete response to the hardening heat-treatment.

Moreover, solution treatment temperature is more important than appears to be shown in the paper; the authors demonstrate that it is possible to widen the range of composition over which an acceptable degree of martensitic transformation, and therefore hardening, is obtained by the use of a lower than usual solution treatment temperature. The converse is also true. High solution treatment temperature or prolonged soaking at above 1100°C can increase the stability of the austenite by dissolving more carbide and render some of the borderline compositions more difficult to transform. In such cases it may be necessary to prolong the treatment at 700°C to overcome the sluggish carbide precipitation resulting from grain growth and more complete solution. In general it is easier to obtain very high strength properties by heat-treatment of cold-rolled sheet and strip than it is in the case of heavy sections, the greater reduction from ingot to the sheet and strip serving to diminish the effects of microsegregation. Fortunately controlled transformation steels have only a limited application for heavy sections as there is generally no point in having a temporary austenitic condition, except for cold forming of thin sections. For bars, castings, and forgings a slight modification has been introduced of the controlled transformation steel known as 520(B) this being 'permanently' martensitic yet having the benefit of very low C content.

Figure *P* shows the results of routine tests on wide coils representing about 50 tons of sheet and strip in a steel 520(S) with 0·05%C, 16%Cr, 5%Ni, 2%Cu, 2%Mo developed by my own company between 1953–56. Referring to this steel in their paper, I notice that the authors have inadvertently omitted the copper content. The composition of the steel is very similar to one of those recommended in the paper, which is perhaps not surprising as the development followed very similar lines.

Although sheets of this steel supplied in the air-cooled 1050°C condition are very readily formed, precautions must be taken to avoid dimensional inaccuracies as a result of a volume change during the subsequent 750°C air-cooling treatment. We have developed a modified heat-treatment which will allow this volume change to be taken up on the strip production line, yet sheets hardened in this manner can be stretch-flattened and some mild forming operations may be carried out. Figure *Q* gives the results of tests on cold-rolled strip, softened in this way for 1 min at 830°C on the production line. Almost complete recrystallization is achieved.

The materials transform during cooling under tension on the production line giving 0·1% proof stress of about 35 tons/in². The resultant sheets may be hardened to give a much higher proof stress by means of a simple 550°C treatment after stretching or forming. After stretching 1% at room temperature followed by treatment at 550°C for about 30 min, a minimum 0·1% proof stress of 60 tons/in² is obtained. Thus it is possible to produce flat stainless steel sheets of 70 tons/in² tensile strength, free from the directional properties usually associated with steels cold rolled to high tensile strength.

Dr **A. R. Entwisle** (Sheffield University): I should like first to ask how the authors measure their M_s. In these martensitic transformations, where the transformation only gradually starts as the temperature is lowered, the M_s will be sensitive to the way in which it is defined, whether 1% or 0·5% is decided upon. What criterion do the authors use? I suggest that for these steels it may be of some advantage to define M_s in rather a different way from the conventional one. What is of interest is the bulk of the transformation, i.e. in the transformation on the more linear part of the curve (Fig.1). It might be more useful to define M_s by the temperature at 20, 30, or 40% transformation, or to extrapolate this linear region back to 0% martensite and get the M_s in that way.

A further point is that with rather lower alloy steels the transformation curves are of this type, and the curved region near M_s is somewhat sensitive to heat-treatment variables. If the criterion for M_s is, say, 0·1% or 1% transformation, there may be some variations for this reason.

I am glad that the authors bring out the point that there is

no one magic number for an alloying element in its effect on M_s, but I was surprised by their curve for tungsten (Fig.12) and would welcome their further comments. Most of their results show a fairly general trend (linear or not departing much from it), but the curve for tungsten is very surprising and I wonder whether there may be some complication of the kind to which I have referred, coming in there.

Some work has been done at Sheffield University on the factors affecting the slope of the curve in the linear range (between about 10% and 60% transformation) and it has been found that it is somewhat sensitive to composition, but not very sensitive, and in the range of composition studied in this paper we should not expect much effect. There is one point, however, which should be mentioned here. Is there any evidence for isothermal transformation in these alloys? Was there any sensitivity of the amount of transformation to the cooling rates? This could be of some practical interest, and it might be better to cool slowly to −70°C, rather than put the steel straight into the refrigerator.

The authors refer to the effect of cobalt on M_s and say that it depresses it. Can they give a figure for this?

In Fig.13(*h*) the hardness for the cobalt-containing steels seems to be low in all cases. The 0% cobalt hardness is lower than in the case of the other steels by 30–40 points. Has there been a slight error in plotting or is there anything significant about this?

Professor **T. Malkiewicz** (Ferrous Metallurgy Research Institute, Gliwice, Poland): In the work done at our Institute we have found a clear dependence of the M_s point on the grain size for Mn and Ni austenitic steels. With coarse grain this temperature was rather higher. Have the authors made a similar observation?

Mr **M. C. T. Bystram** (Murex Welding Processes Ltd): It is likely that some of the controlled transformation compositions will improve the weldability (cracking tendencies) of the martensitic stainless steels, which may be an important aspect of utilization for all types of applications. What are the applications envisaged at present for the controlled-transformation steels and would some of them maintain their high mechanical proportions in prolonged service at temperatures of the order of 400°C?

Written contribution

Mr **K. C. Thomas** (Jones and Laughlin Steel Corporation): The idea of transformation steels goes back many years; in fact, a French patent was issued in 1920[1] for the superficial hardening of austenitic steels by rapid cooling. In 1931, Luerssen and Greene[2] discussed the possibility of obtaining forced-transformation steels by modifying an 18/8 composition and reported some work done in this respect on 0·1–0·18%C–17%Cr–4–9%Ni, 0·05%C–25%Ni–4%Mo, and 0·05%C–25%Ni–0–4%Si alloys. It was shown that a 0·1%C–17%Cr–5%Ni composition had the characteristics of a forced transformation steel, with a tensile strength of 208000/in^2 and an elongation value of 19·8% after sub-zero cooling and tempering at 400°C. It appears that this remained only of academic interest until the 1950's when steels of this type became available commercially. At the present time such steels, i.e. 17/7PH, PH15/7Mo, AM350, and AM355 are used extensively in the US.

The authors state that the M_s–M_f temperature range in the Cr–Ni–Mo and Cr–Ni–Mn steels is about 140°C. As pointed out in the paper the magnitude of this interval will have an effect on the extent of the transformation on sub-zero cooling. It would, therefore, be interesting to know whether variation in the alloy elements affects the extent of this temperature range in the 17/4 base.

The existing mathematical expressions[3,4] relating M_s temperature and composition, show the M_s temperature to be linearly dependent on composition. The results reported for Cu, W, Cr, and possibly Mo do not substantiate this. The results for W and Cu are interesting as they indicate that up to 2% of either of these elements can be added without much depression of the M_s temperature. The paper also stated that all alloy elements are capable of depressing the M_s temperature. However, Co and Al are believed to be capable of raising the M_s temperature when added to 0·2% carbon steels.[5] In the case of the 17/7 base it is known that Al raises the M_s temperature; the presence of about 1·2%Al, in addition to hardening the martensite on subsequent tempering, allows the 17/7PH austenite to transform to martensite on cooling from room temperature to −100°F.

The authors calculated the change in M_s temperature for varying Al contents from Fig.13(*f*) which relates hardness after varying treatments to the Al content. In Fig.11 it will be noted that the amount of δ ferrite increases rapidly with increase in Al content; at 1%Al about 35% δ ferrite is present. It would be of interest, therefore, to know what the effect of Al on the M_s temperature is for fully austenitic structures.

In the case of steels for commercial use, a certain amount of δ ferrite is desirable, up to about 20%, so that a certain amount of grain refinement of the austenite is achieved. It will be noted from the microstructures (Fig.6) that the austenite to martensite change is initiated at grain boundaries leaving patches of untransformed austenite within the grains. We have noted a similar effect in a 0·3%C–12%Cr–6%Ni forced-transformation steel; microhardness measurements confirming the presence of austenite and martensite. If the grain size is decreased in a given composition, then on sub-zero cooling the transformation will be more complete since there will be a larger number of grain boundaries at which transformation may take place.

It would be interesting to know the nitrogen levels encountered in the steels. Ni, being a powerful M_s temperature depressor, has an important effect on the amount of transformation on sub-zero cooling.

In the 17/7PH steel the austenite to martensite change is time-dependent and the recommended refrigeration treatment is 8 h at −100°F.[6] A similar effect, as determined by hardness measurements, has been noted in a 0·3%C–12%Cr–5%Ni forced-transformation steel. Do the authors consider a sub-zero treatment of 2 h sufficient to fully transform the austenite to martensite in their steels?

The forced-transformation steels undoubtedly represent a major contribution in the high-strength stainless steel area. At present they are comparatively costly to produce and very close control has to be kept on the composition and heat-treatment. This is particularly true if a sub-zero treatment is used to transform the austenite. It is likely, however, that less variability in properties may occur if primary tempering is used to effect transformation.

Authors' replies

Dr **K. J. Irvine,** replying on behalf of his co-authors: I agree with Mr Morley about the difficulties likely to be encountered in processing these controlled-transformation stainless steels owing to the pick-up of carbon and nitrogen. This is particularly important when dealing with thin strip or fine wire, since the properties of these steels make them attractive for springs. It is possible to increase the surface content of carbon and nitrogen sufficiently to destroy the controlled transformation, so that considerable care has to be exerted during heat-treatment. Excellent metallurgical work has been carried out by the users who, in dealing with this new type of steel, have experienced many production difficulties which have all been successfully overcome. It has been necessary for them to understand the complete metallurgical background before they could explain what was happening, and alter conditions so that satisfactory properties could be obtained.

Mr Morley mentioned the effect of high solution treatment temperatures. For the majority of steels described in the paper 1050°C is sufficient to dissolve the alloy carbides. However, if strong carbide-forming elements such as niobium or titanium are used then it is necessary to use higher solution treatment temperatures. Reference is also made to the development of modified compositions which in fact transform on cooling to

a martensitic structure. It seems quite certain that these steels will find successful commercial application because they do combine high strength, good impact, and satisfactory corrosion resistance.

In reply to Dr Entwisle, we were mainly interested in the temperature at which the majority of the structure had transformed to martensite and also the temperature at which more than 95% had transformed to martensite. It was found that a satisfactory estimate of the major transformation to martensite by hardness determinations could be got because the 0·08%C martensite has a hardness of about 400 DPN, while the essentially complete martensite transformation could be assessed by a tensile test since it is only with complete martensite transformation that high proof-stress values are obtained. The effect obtained with tungsten was unexpected but has been reproduced; further work will be necessary to explain this. We did obtain an estimate of the effect of cobalt in depressing the M_s, and over the range of cobalt contents studied, the effect was that 1%Co depressed the M_s by about 10°C. There was nothing unusual about the hardness values of these martensitic structures containing cobalt. On looking again at Fig.13(*h*) it would appear that this diagram has been misplaced by one division on the graph in setting up the block so that the hardness values are in fact, 50 DPN low. The average hardness for these cobalt steels was 420 DPN.

There were several questions concerning the effect of coarse grain size on the M_s. The range of solution temperatures used in this work was not very large and was, in general, 950–1150°C. Over this range there was not much variation in grain size and no apparent effect on the M_s.

Mr Bystram asked about high-temperature properties of these steels. It should be realized that the alloy contents of the majority of these steels were designed to give a particular combination of structure and M_s. The properties of the steels were thus of tempered martensite. There is, of course, a large amount of background information on the ways in which tempering resistance, and hence high-temperature properties, can be increased and, of course, it is possible to modify these steels in the same way. Increased additions of Co, Mo, and W, for instance, improve these properties. It is necessary, however, to modify the base composition so that the controlled-transformation characteristics can be maintained.

The written contribution from Mr Thomas contained some interesting information, and by the reference to earlier work it showed the difference between laboratory work and developing a successful commercial steel. There is a question about the effect of alloying elements on the martensite transformation range. As stated in the paper this range is about 140°C and the element which affects this range the most is nickel. Increasing nickel contents considerably lengthen the temperature range for the last 10% of martensite transformation. It is difficult to assess this effect exactly because of the difficulty of measuring the M_f. However, it does not appear that other alloying elements have such a marked effect. It should be remembered that there is a possibility of local segregation and also the complication of alloy differences between austenite and delta ferrite, which can exaggerate this effect. The evidence is that cobalt and aluminium depress the M_s temperature and it is considered that if the reverse appears to be the case with an element like Al, then it may simply be due to the fact that the combination with nitrogen reduces the effective nitrogen content. It should be emphasized that the steels used in this work were all low-nitrogen steels. The content aimed for was 0·02% and the steelmaking was carefully supervised to ensure that this level was obtained. When making small ingots this does require careful control both with regard to the ferro-chromium used, and the necessity to make wash heats in between each cast. Otherwise the high-chromium skull remaining picked up a large amount of nitrogen.

In these low-carbon steels there has been little evidence of isothermal transformation of martensite, and consequently it was found that 2 h at the refrigeration temperature was sufficient. No further increase was obtained with considerably extended refrigeration times.

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Joint discussion on fuming in steelmaking furnaces

This discussion was based on the following papers (dates of publication in the *Journal* are given in parenthesis):

- **'Fumes from the acid open-hearth process'**, by *W. Strauss* and *M. W. Thring* (1959, **193**, 216–221)
- **'Preliminary studies on the evolution of fumes from iron at high temperatures'**, by *E. T. Turkdogan* and *L. E. Leake* (1959, **192**, 162–170)
- **'Factors affecting fuming in open-hearth furnaces'**, by *C. Holden* (1959, **193**, 93–102)
- **'Prevention of brown ferric oxide fumes by steam blowing'**, by *H. Kosmider*, *H. Neuhaus*, and *P. E. Hardt* (1959, **193**, 6–11)
- **'Some factors affecting fume evolution from molten steel during oxygen injection'**, by *K. Knaggs* and *J. M. Slater* (1959, **193**, 211–216).

Professor **M. W. Thring** (University of Sheffield), in presenting the paper by himself and Mr W. Strauss, said: The present type of OH furnace, from the point of view of fuming, is unsatisfactory. It has a slag pocket which, with luck, may remove some coarse material, regenerators which may take out a little more and generally block up in so doing, and finally, again with luck, there is some kind of filter attached at the end. We felt that the OH furnace was not really designed for dealing with oxygen-blowing processes and that we had better consider how this could be remedied. We decided to combine the regenerative system for air pre-heat with fume removal, and instead of a coarse-packed regenerator with a loose packing of bricks, to have a comparatively finely packed regenerator, removing the particles there, on the same principle as certain types of filter.

If this fume-removal filter is combined with a regenerator, air is blowing through it alternately with the waste gases, and therefore when the waste gases are going through it the brick surfaces are very much colder than the gases. Because of this temperature difference there is a considerable thermal precipi-

tation effect and it is possible to achieve 90% efficiency in fume removal using granulated material with an extended surface. How long it can be run before it blocks up and whether or not we should have a filter consisting of coarse bricks and then finer material with very fine material at the bottom, are questions which remain to be studied. This paper gives the information which Strauss obtained in the coarse of his work in testing a combined regenerator and fume-removal filter; Fig.1 shows where he took his samples from the regenerator. At the point indicated 'To suction fan' he had his experimental cleaning and regenerative system, alternately blowing and sucking the gases through in the same sort of cycle as with a regenerator.

The facts which Strauss found have been published in this paper. During oxygen lancing he sometimes obtained the very high concentration of 6 g/m³ of fume; the Ringelman smoke number for the top of the stack during that period would be about 4, which, if it were black, would be called a serious smoke emission, but during the rest of the process and in refining, when not working with oxygen, there was a maximum Ringelman number of 2 which could be ignored.

He found that the fume contained many other things besides iron oxide. From Table IV it will be seen that the iron-oxide content varied in different samples from about 12% up to nearly 90%, so that the material was by no means 100% iron oxide. That part of it which was iron oxide was a mixture of hematite and magnetite. The particle size of the fume was of two types, but the main bulk of it was very small and below 1 μ in size.

If metallurgists cannot obtain the benefits of oxygen lancing without fume emission I think that we can probably find a solution to the fume-emission problem on the lines which I have suggested.

Dr **S. Klemantaski** (BISRA) presented the paper by Dr Turkdogan and Mr Leake, Dr Turkdogan being in the USA.

Mr **C. Holden** (The United Steel Companies Ltd) presented his paper and showed a short film illustrating the effect of carbon in the melt on the quantity of fume evolved.

Dr **H. Neuhaus** (Klöckner-Werke AG, Hütte, Haspe) presented the paper by himself, Dr Kosmider, and Dr Hardt.

Mr **K. Knaggs** and Mr **J. M. Slater** (BSCRA) presented their paper and showed a film of the inside of the furnace during oxygen injection.

Mr **B. Larsen** (Edgar C. Bain Laboratory for Fundamental Research, United States Steel Corporation) then showed a film entitled 'Bubbles in Steelmaking'.

Mr **B. M. Larsen:** One receives two general impressions from most of the recent papers on iron fume formation. First, that there are at least three modes of fume development in commercial iron or steel refining processes: (*a*) simple vaporization occurs as an approach-to-equilibrium effect wherever liquid metal contacts a gas phase, (*b*) partial or complete 'burning' of iron droplets occurs, probably producing some fumes whenever these are thrown up into an oxidizing gas phase by the bubble-bursting effects in the boil processes, and (*c*) neither of these effects seems adequate to explain the bulk of the total fume formation, so that, seemingly, some 'X-mechanism', the true nature of which is not yet understood, must also occur at least in periods of heavy fume development.

The other impression is that with this complexity, we can probably easily misinterpret observations made in furnace operations. For example, visual observations strongly suggest a very effective 'washing out' of fume from boil gases as these pass through a slag layer, especially in foamy slags. It would seem that such effects must occur to some extent. The X-mechanism might somehow form a concentration of iron fume inside a bubble which was much above the metal vapour pressure at the existing ambient temperature. As such a bubble moved through more liquid metal, the excess fume could condense back into the liquid metal; or, in a foamy slag, any fume present as oxide particles could simply be absorbed into the slag, or if present as metallic vapour or liquid droplets, it could react with the slag, as for example by the reaction

$$Fe + Fe_2O_3 \longleftrightarrow 3\ FeO$$

which at moderate slag temperatures might almost completely remove iron fume, giving boil periods when apparently pure white issuing gases are observed. We may also misinterpret the issuance of gases nearly free of iron oxide as being largely due to such a washing-out effect, whereas in reality it may simply indicate the absence of certain special conditions requisite for much excess fume development by the 'X-mechanism'.

In this situation, admitting the primary need for more experiments, it may also be reasonable to indulge in some speculation as to the reality and the nature of such an X-mechanism. Assume first, even though this assumption may not seem consistent with all the effects observed, that this mechanism depends on the contact of a more or less highly oxidizing gas phase with a liquid–metal surface rather high in carbon content. Two things must be happening at the same time: (*a*) a rapid absorption of oxygen at the liquid surface, and (*b*) formation of CO molecules at this same surface so that a steep gradient in carbon concentration in a thin liquid-diffusion film layer produces a rapid diffusion of carbon atoms to the interface. If we also assume that, at least in a certain range of temperature, a very thin film of liquid wüstite tends to form on the liquid surface, it seems quite reasonable that the interaction of carbon atoms from beneath this film should cause a simultaneous formation of a CO molecule together with the rapid freeing of the linkages of one Fe atom so that it could easily leave the liquid surface. This would be a process occurring in such a steep activity gradient as to be somewhat like an explosive reaction confined to an interface, with several apparent advantages. It would not need the assumption of any complexes such as FeCO; since it could be affected by separate kinetic factors, it might under most favourable conditions approach an apparent stoichiometric relationship with one Fe atom released for each CO molecule evolved, yet, with a different balance between transport rates, most of the Fe atoms might relax back into the liquid phase. The rate of fume formation would naturally be low at low carbon levels since the adsorbed oxygen could diffuse into the liquid faster than the rate of supply of carbon atoms so that many CO molecules might form separately with much less disruption of any FeO-film layer formed at the interface. A certain temperature range could also be optimum, since many observations of filming-over of spoon samples indicate that higher superheat and perhaps lower liquid viscosity favour the absence of the oxide film formation assumed here to be required for the very rapid 'loosening' of Fe atoms into the gas phase. Such a mechanism combined with a variable 'washing out' of fume before leaving the bath might explain in part the variable fume evolution in Bessemer operations. Fume formed during oxidation of phosphorus at a high rate could be caused by a mechanism more or less analogous but different in detail.

As to further experimental work, the most immediate need is for further careful measurements of fuming rate *v.* temperature at constant conditions of gas flow and composition and carbon content in the liquid iron.

Professor **F. D. Richardson** (Imperial College of Science and Technology): There are many puzzling facts in the five very interesting papers on fume formation: it is not easy to pick out what is really important and what is incidental. It seems clear, however, that carbon and oxygen are very important, and that temperature may be. To get a proper perspective, it is helpful to compare the true vapour pressures of iron with the vapour pressures which would be equivalent to the quantities of fume that have been observed.

The vapour pressure of iron is about 2×10^{-5} atm at 1 500°C, and 5×10^{-3} atm at 2 000°C. If the laboratory results reported by Turkdogan and Leake are considered, the iron pressure equivalent to the fume is 10^{-1} atm, a figure much higher than most of the values calculable from the other papers. It would be most interesting to know whether such a

high value was observed by Mr Holden in the laboratory experiment shown in his film; a very dense fume was clearly coming from the high-carbon iron which he was treating at about 1100°C.

In the work of Knaggs and Slater, we find equivalent pressures of about 5×10^{-2} atm, and in the OH work values of the order of 2×10^{-3} atm. All these pressures are so much higher than the true vapour pressures, even at 2000°C, that simple vaporization at local hot spots cannot be an adequate explanation.

So far as converters are concerned, the general idea seems to be that very high local temperatures are probably the biggest single cause of intense fuming in the presence of oxygen. I have difficulty in believing that this accounts for everything. As far as I can see, the high local temperatures have been derived from calculations that take no account of heat transfer away from the rising gas bubble, i.e. heat transfer from the surface of the bubble, which becomes coated with iron oxide, into the body of the melt. If reaction between gas and metal is to proceed, iron must diffuse through the oxide film. This mass transfer of iron must be very much slower than the conduction of heat away from the bubble into the body of the melt, for thermal diffusivities are many times greater than diffusion coefficients. What is the basis for the assertion that there are measured temperatures which support these incredibly high calculated values? I assume that they must be temperatures measured through the bottom of a tuyère. My recollection of IRSID's studies with two-colour pyrometers is that the observed temperatures were not particularly high; and these surface temperatures would probably fall as the bubbles rise through the melt.

There is a further objection to an explanation based solely on temperature: it cannot account for the effect of carbon which is evident in many cases, and particularly in the results given in Table II of Knaggs and Slater. A well controlled series of temperature measurements with bottom blowing with oxygen or oxygen–helium mixtures, in the absence of carbon, would be of very great help. A better idea of the maximum effects which we can attribute to temperature could then be obtained.

The indubitable effect of carbon makes me incline strongly to the view that there must be some chemical explanation of these phenomena. The reduction of fume brought about by adding steam to the oxygen, hitherto attributed to a lowering of temperature, might also be caused by chemical effects. If we examine the compositions of the top gases as reported in the paper by Kosmider, Neuhaus, and Kratzenstein (1954), we see that the $CO:CO_2$ ratio is very much greater when steam is added; there is a greater efficiency in the utilization of the gases in the bubble, probably caused by some catalytic effect or faster diffusion rates with hydrogen and steam. The introduction of steam thus brings about substantial chemical changes and it may be these rather than the lowering of temperature, that are the cause of fume suppression. It would be helpful therefore if Dr Neuhaus could say whether or not the curves which were published for the composition of the emergent gases by Kosmider *et al.* in 1954 are typical. Can we still take them as correct? I should like to know also how they were measured and how far away from the top of the melt they were recorded.

In inclining to chemical theories, I do not want to suggest that I am inclining to Kohlmeyer's carbonyl theory. The theory is conceivably true but the work on which it is based is very crude and preliminary, and I think that there are other possibilities. Mr Belton, working with me with the support of the United Steel Companies Ltd, has been studying the vaporization of iron in the presence of the gas mixtures of hydrogen + water and carbon monoxide + carbon dioxide. We now feel sure that iron is volatile as a hydroxide, possibly also as an oxide, and we are finding that these volatiles can exist in the gaseous state at 1400–1500°C at partial pressures which are some 20 times greater than the true vapour pressures of iron. This sounds a big effect, but the pressure of these volatiles at 1500°C is equivalent to no more than the true vapour pressure at a temperature 200°C higher. It seems unlikely that these volatiles could come anywhere near to accounting for the equivalent pressure of 10^{-1} atm obtained by Turkdogan and Leake: they could however account for the OH observations.

Since we are still in difficulty, we should perhaps think about bursting bubbles and the material which is thrown into the gases whenever a bubble bursts. It seems to me that in laboratory experiments the effect of bubble burst may be much more important than in practice, because there are shorter times and distances in which particles can settle out of the flowing gases. This phenomenon might thus account for the rather high equivalent pressures observed by Turkdogan and Leake. It is only when carbon monoxide is coming off that bubbles can form and burst in their experiments, and it is then that they get their big effects.

I should like to put a question to Mr Holden. He showed us two very impressive photographs of fuming during the injection of oxygen in the open hearth, with the bulk of the fume lying like a cloud above the bath and an apparently clear zone in between. I think he said that all the samples of gas and fume abstracted from this clear zone yielded particles of unoxidized metal. It would be interesting if he would say how these samples were taken and whether the quantities of material which could be abstracted from this clear zone were substantial. Could it be that in this clear zone there are present volatile iron compounds that decompose on cooling and that the dense fume above is produced when these compounds are oxidized to yield very finely divided solid material?

Dr **H. Neuhaus** (Klöckner–Werke AG), in reply: Professor Richardson asked whether the curves for the waste gases in our earlier publication are typical. In our opinion they are. We obtained a large number of gas samples inside the converter and in the chimney. Only the average values are shown in the curves. The samples inside the converter were taken about 1500 mm above the level of the bath.

Referring to the main question whether the brown fume is produced by vaporization or not I should like to say: 'Yes it is in the basic Bessemer practice'.

Our own experimental work relates only to the bottom-blowing converter process. The colour pictures that have been shown indicate that there is indeed a fume evolution during decarburization, but the heaviest fume is emitted during dephosphorization when the bath is free of carbon.

Because of the submicroscopic grain size of the Fe_2O_3 in the dust we suspected some years ago a vaporization effect. It is quite impossible to calculate the correct temperature of the hot spot, but we found for different refining gases a correspondence between the roughly calculated values of local temperature and the fume emission. Some years later Naeser *et al.*[1] and von Bogdandy[2] measured independently the hot-spot temperature through the bottom tuyères by two-colour pyrometer and found similar values.

Whether our temperature is correct or not is not so important, but we believe it indicates a correct tendency and helps us to progress in this important matter.

It may be that in OH practice another mechanism of fume formation occurs, particularly during decarburization. This second possibility seems more to be a chemical effect, perhaps interactions between the Fe, the oxygen, and the carbon.

Nevertheless, further investigation may construct a bridge between both theories.

The Chairman, Dr **J. H. Chesters** (The United Steel Companies Ltd): Do you think that temperatures of the kind about which we are talking come into the OH problem with which Mr Holden has dealt? When oxygen lancing is not used, presumably the temperatures must be lower. Do you think that both the OH (without oxygen) and the basic Bessemer converter are to be explained in the same terms, or are they different effects?

Dr **Neuhaus:** I think that they are two different effects.

Mr **C. Holden,** in reply: Professor Richardson asked whether or not the vapour pressure of the fume was measured during the laboratory experiment shown in the film. The answer is that they were not, because I have never been convinced that it is possible in the laboratory to simulate anything approaching what is happening in the OH furnace. We did, however, obtain some very carefully made samples of iron with various elements (C, P, Si, and Mn), added to them and the effect of these on fuming has been examined by Mr Barnsley. The results, I believe, were very similar to those obtained by Turkdogan and Leake. Whereas values of vapour pressure might be interesting from a process-chemistry point of view, the paper which I presented was intended more as a practical man's survey of the situation, and I doubt whether it is worthwhile to go further than this from the point of view of the OH operator.

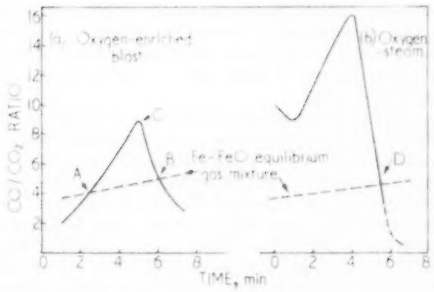
With regard to Fig.8 in my paper, we have always noticed the clear zone to which Professor Richardson referred. Up to a few months ago I had never thought deeply about it, but our attention was concentrated on it when we heard from Knaggs and Slater that they had in fact sampled pure iron from this zone (they give the results in their paper) during oxygen lancing in an arc furnace. More recently Mr Pengelly of the United Steel Companies, has taken similar samples from arc furnaces and has confirmed that finding. The difficulty when drawing samples from the bath is to quench them adequately, and because the gases must contain a fair amount of hot CO it could be possible to sample iron oxide and reduce it to pure iron within the probe.

Mr **A. E. Pengelly** (The United Steel Companies Ltd): These experiments were undertaken to test the water-cooling of a sampling probe designed to extract gas and solid samples from the atmosphere above an arc-furnace bath and to quench them to below 300°C. Samples obtained were incidental to the testing and no careful measurement of gas flow was therefore made.

Four samples were taken from an arc furnace at Samuel Fox's works, three of them during oxygen blowing. The first was a complete failure owing to poor probe positioning. The second was fairly successful, but owing to a billowing brown fume cloud, probe positioning was still difficult and the filtering nose spent some proportion of time in the cloud. This sample contained about 70% of alpha iron, the rest being a mixture of FeO and Fe_2O_3. The total weight in the filter was less than 1 g, but substantially more than $\frac{1}{2}$ g. The third sample was taken with the probe properly held in the clear fume-free zone above the bath and contained only alpha iron. Unfortunately early in this sampling period, the nozzle became blocked by slag, so that the total weight was only 50 mg. The fourth sample was taken after oxygen blowing and contained only zinc oxide.

I made some rough calculations of the effect of iron vapour pressure present during extraction of the second sample. Assuming that the iron was all vapour, its effective partial pressure within a factor of 3 or 4 was 4×10^{-3} atm. It is difficult to estimate the local bath temperature, but the average was certainly nowhere near 2000°C.

Mr **G. R. Belton** (Imperial College of Science and Technology): We are trying, as Professor Richardson has said, to find the fundamental mechanism of fuming, the research being sponsored by the United Steel Companies. Some years ago Miss M. Gleiser[3] at Imperial College carried out measurements on the solubility of oxygen in iron and used water-vapour hydrogen mixtures to give the desired oxygen potentials. She found a weight loss of iron which could not be attributed entirely to the vapour pressure of iron, so we took this as the starting point of the research. The hydrogen-water vapour mixtures were obtained by bubbling a metered volume of hydrogen through a pre-saturator and condensing out the excess water by passing through a glass vessel, filled with glass beads, held at the exact saturation temperature in a thermostatically controlled water bath. In order to stop condensation, all the glass



R *$CO:CO_2$ ratios of exit gases during Bessemer blow, taken from Kosmider* et al.

tubing through which the gas mixture passed was heated to about 100°C by means of resistance wire wound spirally on the tubes. Measurements of the vapour pressure of iron were then carried out in this atmosphere by the transportation technique. An enhanced volatility has been found but at this stage it is not possible to distinguish whether the volatile species is a hydroxide or an oxide of iron.

If the volatile species proves to be an oxide and it is stable when the oxygen pressure is below that of the Fe–FeO system it will explain some of the observed phenomena. I have taken the gas mixtures observed by Kosmider, Neuhaus, and Kratzenstein[4] when they used enriched oxygen in the standard basic Bessemer, and I have plotted the $CO:CO_2$ ratios against the time of blowing. If we draw on this a line representing the equilibrium gas mixture for the Fe–FeO system we obtain Fig.*R*.

For a volatile oxide unstable at oxygen pressures above that of the Fe–FeO equilibrium, maximum fuming would be expected to occur at points *a* and *b* in Fig.*R* and some reduction in fuming would be expected at the point where the gases are most reducing (point *c*). If Fig.2(*a*) in the paper under discussion by Kosmider, Neuhaus, and Hardt, is examined, it can be seen that there is this tendency towards two distinct fuming peaks.

Coming to what is to me a far more interesting result, I calculated the effect for the steam-blown converter. Here again, the $CO:CO_2$ ratio for the Fe–FeO system can be plotted, and we find that the gas is always strongly reducing towards FeO and becomes oxidizing only towards the end of the blow as shown in Fig.2. Reference to Figs.3 and 4 in the paper by Kosmider *et al.* reveals, at first, little fuming and then a period of fuming towards the end, corresponding to point *d* in Fig.*R*.

Leroy and Simon[5] have reported an almost complete absence of fume in the IRSID continuous desiliconizing process. This, I think, is in favour of our mechanism. The porous refractory bottom of the lander splits the oxygen stream into very fine bubbles, so we would expect the gas mixture to be very near equilibrium composition. The equilibrium gas mixture for iron with, say, 2% carbon, contains a far larger proportion of CO than any mixture on my figure,[6] so we are far removed from the optimum conditions for fuming.

This is only a tentative mechanism which we are putting forward and may have to be revised if we find any differences in the suggested behaviour of the volatile compound, but it lends a little support to the chemical approach.

Dr **J. Pearson** (BISRA): The whole picture presented by these five papers is one of certain contradictions and confusions, but also of some general agreement not only between these authors but with what we have read in the past. First of all, we find that in the laboratory experiments carbon is necessary to get fuming, but in connexion with the maximum period of fuming in bottom-blown converters carbon is not necessary, so that it seems that there must be two mechanisms at work. It was also found that in most of the laboratory experiments so far described carbon had been removed at a fair rate during the fuming, but in the experiments which Knaggs and Slater describe

when they had a high silicon content they obtained measurable fumes but no change in the carbon content. This may be because the amount of fume was very small and if it had involved some carbon it could not be measured in terms of analysis of the steel.

One thing seems clear about the work reported by Knaggs and Slater and now by Belton, and also confirmed by Turkdogan and Leake: immediately above the bath, before air or oxygen has access to the fumes, the metal is at least predominantly iron, and therefore we can assume that there was vaporization. We have also learned that temperature is important and that the higher the temperature the greater the amount of fume. This is not surprising, since presumably the production of fumes involves a physical or chemical mechanism, and a physical or chemical mechanism generally accelerates with increase of temperature. The curve of vapour pressure which appears in the paper by Kosmider, Neuhaus, and Hardt is of the same shape as that for the rate of reaction at increasing temperature.

In connexion with some of the figures which Professor Richardson has given and in relation to the amounts of fume evolved, it may be of interest to repeat something that I said at the 47th Steelmaking Conference when I was reviewing this question of the formation of fume and mentioned the work of Kosmider and Neuhaus. I said that I thought it unlikely that the temperatures which they calculated for the hot spot would persist throughout the bath, but I assumed that a bubble, if it contained any vapour, would leave the bath at the temperature prevailing at the time the blow was conducted. I then worked out from the curve of the vapour pressure of iron as a function of temperature what should have been the vapour pressure of iron in bubbles leaving the surface, and from the figures which Kosmider *et al.* gave of the volumes of waste gas etc. I calculated what should have been the iron-oxide content in the gas which they measured. My calculations gave a figure of 0·8 g/m³ and the actual figure determined was 0·4–0·9 g/m³. In other words, there was close agreement between the amount of iron oxide found by Kosmider *et al.* and that derived from the vapour pressure of iron at a temperature of 1 580–1 600°C.

I should like to ask a question about the effect of water vapour on the suppression of fuming. It is in fact feasible to assume that the dissociation of water vapour cools the hot spot, thereby reducing the local temperature, and that this has an effect on the vapour pressure and less fume is evolved. The effect is clear not only in the film which Dr Neuhaus showed but in experiments which have been carried out in various parts of this country; but, and here I am putting a question to those who have more experience than I have, CO_2 is also added to bottom-blown converter heats for precisely the same reason that steam is added, namely to cool, and the amount of CO_2 added to the oxygen is measured so as to have the same effect as the water; yet I am told that an oxygen–CO_2 blow is particularly obnoxious from the point of view of fume production. If this is true, then CO_2 as coolant is doing nothing to suppress fume formation, but it may have some connexion with what Mr Belton has told us.

The Chairman, Dr **J. H. Chesters:** This difference between the OH furnace and the converter is very puzzling. It is possible to explain most of what is seen in the converter by very high temperature, but Mr Holden has found, and other workers in the United Steel Companies have repeatedly found, that the fuming does not coincide with the maximum temperature in the furnace. When we put cold scrap in and heat up with pitch-creosote and watch the roof temperature, there are two peaks inside 10 min and they occur well before the roof temperature gets to the maximum. One possible explanation is that the bath becomes coated or the scrap becomes coated with slag. Would Mr Larsen's bubble mechanism explain why the peak comes at a lower temperature than the maximum?

Monsieur **B. Trentini** (IRSID, Paris): In the bottom-blown converter, it is well known that the higher the oxygen concentration in the air which is blown, the more fume results. In the last part of the blow, at a given percentage of the refining, less fume results if we blow with normal air than with 40% oxygen in the air for the same bath temperature.

Another fact should be noted. In a certain plant in France we use different types of bottom in basic Bessemer converters. We have one type of bottom (*A*) with a uniform distribution of holes in the bottom. Another type (*B*) with groups of holes in magnesia blocks: the dolomite has no holes and the magnesia has a series of about seven holes very close together, so that the concentration of holes in that area is much larger than in a bottom of the first type. It is well known to engineers of this plant that for a given temperature this second type of bottom, *B*, gives much darker and stronger fume than the first type, *A*.

This is also important in considering fume formation. I was told by Monsieur Decamps from Sollac that when the holes get larger towards the end of the life of a bottom, the amount of fume increases very quickly and the behaviour is very different from that with a normal bottom. This phenomenon is used as a check-up by melters.

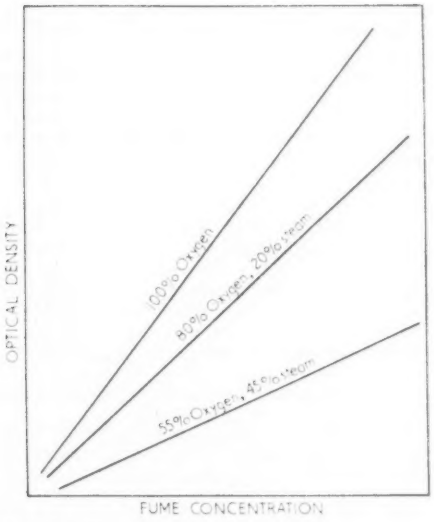
As far as top blowing is concerned, if we use, for example, oxygen with a high concentration of lime in it, we can suppress the fume entirely. We have made experiments in an OH furnace and find that if we blow oxygen alone with a given carbon in the bath, around 0·5%, there is at the start a great amount of red fume. A few seconds later, if we add lime powder in high concentration, about 12 kg/m³, we suppress almost entirely the red fume. We can explain that by the effect of local lower temperature.

There is a third point which I want to emphasize which concerns what Mr Larsen said about a foaming slag and the lowering of the fume when there is a layer of slag above the metal. We have noted that this is very marked with the top-blowing process. When there is little or no slag cover the amount of fume is very great, but with a thick layer of slag the amount of fume, more or less foaming because of high carbon evolution, is much less. It seems, therefore, that slag acts as a filter. On that same point, in the case of the basic Bessemer process we have developed a method of measuring the opacity at the end of the blow and have found that it decreases. We think that that is because the slag in the basic converter is shaped only at the very end of the blow, in the last few seconds, and we believe that when that slag is formed, it filters a part of the fume.

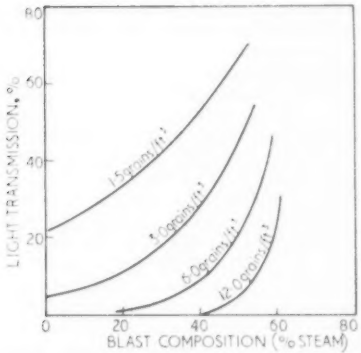
Mr **R. W. Evans** (Steel Company of Wales Ltd): Originally I thought that an oxygen jet impinging on the surface of molten steel increased the temperature to a point where iron-oxide vapour was generated, which accounted for the fume. Considering the bottom-blown Bessemer process, however, it now seems to me impossible that the high temperature locally generated by a blast containing oxygen under the metal should persist while the bubbles are rising through the metal. If the high temperature did persist then the temperature in the area of the bubble lifting itself through the metal surface would be higher than the average temperature of the metal, but, bearing in mind the opportunity for equalization of temperature while the bubble is rising, this seems extremely unlikely.

One should consider, also, the effect of fume on steam addition to the blast in the bottom-blown converter. As the carbon is exhausted in the bath, the gas leaving the converter progressively becomes richer in pure hydrogen; towards the end of the blow when the carbon is exhausted almost pure hydrogen is being evolved. There is dissociation of H_2O; the oxygen is used in oxidizing the metal and hydrogen comes off. This may be monatomic or ordinary hydrogen, but pure hydrogen is a very strong reducing agent. Can this have an effect on the constitution of the iron oxide so that it seems to be almost colourless? It may mean that much the same quantity of material as before is coming off, but it does not look so bad.

Further evidence of this possibility is that although fumes from OH furnaces and Bessemer converters may both be dense when oxygen is used, there is a difference in the colour



S *Effect of steam on optical density of fume*



T *Effect of steam on light transmission at different fume concentrations, calculated from Fig.S*

tint. The OH fumes have some resemblance to powdered hematite with a definitely red tint, whereas the fumes from the converter are definitely brown.

It is also possible, when using roof lances on an OH furnace, to reduce fume considerably by the addition of steam, and the same effect occurs when desiliconizing hot metal. Perhaps this is a chemical effect associated with the evolution of a reducing gas.

Dr **S. Klemantaski** (BISRA): The apparently excessive vapour pressure referred to by Professor Richardson might be due to (*a*) mechanical iron carry-over (by droplets) in addition to that due to evaporation, (*b*) the true iron temperature being higher (through local oxidation of iron and/or carbon monoxide) than that measured, and (*c*) the production of droplets so small that their vapour pressure was considerably greater than that over a relatively flat liquid-iron surface at the same temperature. Combinations of these effects are possible and their direct experimental detection would be difficult.

I should also like to refer to the temperature of a gas bubble passing through liquid iron. Little convective heat transfer can be expected between gas and metal, nor could radiation account for much (because of the poor emissivity of the main component, CO): thus a bubble which starts appreciably hotter than the bulk of the metal in a converter will probably stay so when it breaks surface and will thus be able to convey concentrations of iron vapour greater than expected from metal-temperature measurements. If larger holes in the converter bottom mean larger bubbles (and hence less convective heat transfer) the case mentioned by Monsieur Trentini furnishes some support for this view. The relatively high thermal conductivity of hydrogen would make the temperature of the bubbles containing this gas closer to that of the metal, and this may be a factor in fume reduction by steam additions to the oxygen.

Dr **J. Pearson** (BISRA): Mr Larsen made a point which Monsieur Trentini was able to confirm from practical experience. I should like to quote a case where, apparently, operating practice and laboratory experiments seemed to fit in with one another. In the particular field which we are discussing they do not do so. I have to refer to the work of others, and in particular earlier work by Dr Turkdogan. At that time he had a furnace which consisted essentially of a pot with iron, containing carbon, in it, into which he blew oxygen through a narrow jet. Initially he placed on this iron lumpy slag-making material and blew oxygen through the jet, and he obtained copious fumes. When the slag had fluxed down and the reaction was continuing satisfactorily the slag would come to the top and stay there. After a short time he would stop blowing to take a sample of the metal, the slag then forming a thin layer. He then blew again and the slag parted and there was a lot of fume, but then the slag would foam and the fume would stop. That is an example of how, in the laboratory, slag was acting to filter out fume.

Professor **F. D. Richardson** (Imperial College): I think that Monsieur Trentini could give us an idea of whether the temperature with oxygen blowing, determined by measurement by an optical pyrometer viewing up the tuyère, is in fact very different from that observed with blowing by air. Is it a matter of 1000° or 500° or 100°C?

Monsieur **B. Trentini** (IRSID, Paris), in reply: We have carried out experiments with a pyrometer looking through the bottom of a basic Bessemer converter and have used it to find the evolution of temperature during the heat. We took a normal hole or tuyère in the bottom of the converter and had there our two-colour pyrometer and obtained the normal curve of temperature evolution in a basic Bessemer converter.

We blew a small amount of nitrogen to keep the hole open (but not too much, to avoid cooling it down), and we could check the temperature by the immersion pyrometer to be sure that we had the right temperature. At the end of this series of experiments we blew nitrogen-air or enriched air. As might be expected, much higher temperature at the impact zone was obtained with oxygen-enriched air. Similar experiments have also been conducted in Germany.

Mr **K. H. Sargent** (British Oxygen Research and Development Ltd): Since Mr Evans has asked whether the addition of steam alters the nature of fume, as distinct from its weight, I should like to present some evidence obtained at BORAD. In these experiments, 1-cwt batches of hematite iron were desiliconized by top blowing. Extending the conclusions to a larger scale, to decarburization, or to bottom blowing, requires careful consideration.

Two properties of the fume were measured: its weight concentration and its optical density. These two properties are linearly related, but the exact relationship depends upon the blast composition and conditions of blowing. Figure *S*, for example shows that for a given concentration of fume, that generated by oxygen/steam has a lower optical density than that generated by oxygen. Figure *T* translates these results into more visual terms, and shows that at high concentrations of fume, a sudden increase in light transmission can be expected in the region of 50–60% steam.

We do not believe that this change in obscuring power completely explains the value of steam, but it does prove that steam can alter the nature of the fume. This is supported qualitatively by a change in colour of fume samples from brown to black.

TABLE 1 Results of oxygen jetting on iron-base alloys

Alloy	Analysis, % Main alloying elements C	Si	Mn	P	Observed effect	Relative amounts of fume evolved
1	4·0				Dense orange coloured fumes	1·0
2	3·65	1·15			Reddish brown fumes, vigorous reaction	1·2
3	3·58	2·90				2·3
4	0·91	16·45			No fumes observed	...
5	3·82		2·7		Dense brown fumes, containing a black deposit	3·4
6	4·15		15·86			4·1
7	3·60			0·86	Dark brown fumes, violent reaction	5·4
8	3·38			2·60		3·2
9	0·87			15·70	Grey fumes tinged with brown	0·4
10	3·70		1·12	1·01	Dense reddish-brown fumes which changed to dark brown	3·6
11	0·05			0·58	No fumes observed	...

The optical results suggested that the particle-size distribution had been changed, possibly by the formation of a hydrogen flame and consequent retarded nucleation, as proposed for the Rotor.[7] Some preliminary particle-size analyses, in fact, show a broadening of the distribution with a distinct decrease in the cross-sectional area per unit weight. Alternative suggestions were: (*a*) a chemical effect of the hydrogen on the fume, and (*b*) a localized cooling of the reaction zone. We tested these hypotheses by injecting nitrogen and hydrogen as alternative diluents with the oxygen, and obtained surprising results. Nitrogen behaved very like steam in depressing the optical density/concentration graph with only a slight suppression of fume weight. Hydrogen, on the other hand, had little effect on the optical density/concentration graph, but produced a substantial reduction in fume weight. It appears, therefore, that the optical effect of steam is due to cooling of the reaction zone, and this is supported by the results of Trentini and Allard on the injection of powdered lime.[8]

Finally, it should be mentioned that although we have extended our observations on fume suppression to the use of inflammable gases other than hydrogen, e.g. coal-gas and propane, and to decarburization, the efficiency of suppression (60–80%) is not yet of commercial interest. If it could be developed to a level acceptable to the Alkali Inspectorate, however, it might have the advantage over steam of avoiding the temperature loss inevitable with steam.

Written contributions

B. P. Barnsley (The United Steel Companies Ltd): Mr C. Holden mentioned tests which have been carried out by the United Steel Companies Ltd, regarding fume evolution during oxygen lancing on a variety of alloys. Several selected iron-base alloys were melted in a laboratory furnace and then subjected to oxygen jetting. The alloys generally contained about 4% carbon and varying amounts of silicon, manganese, and phosphorus. The results were mainly quantitative, but rough measurements of fume emitted in a given time were also made. These indicated that the presence of Si, Mg, and P further increased the amount of iron oxide evolved (*see* Table 1).

Dr **H. Neuhaus,** in a written reply: Referring to Dr Pearson's questions about fume formation with oxygen–carbon dioxide blows, it may be that there is a difference in the velocity of reaction between the endothermic reaction of vapour with Fe and of the carbon dioxide with Fe.

I cannot agree with Mr Evans. It is true that towards the end of the blow hydrogen and steam are being evolved out of the converter mouth. Hydrogen is indeed a strong reducing agent, but on the top of the chimney all waste gas, and the iron vapour too, must be oxidized.

Furthermore, it is a fact that the quantity of fume particles in steam blows is only ten times less than in normal air blows. Therefore the effect of steam addition to oxygen is, in our opinion, a physical effect.

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Joint discussion on tempering steels

This discussion was based on the following papers (dates of publication in the *Journal* are given in parentheses):

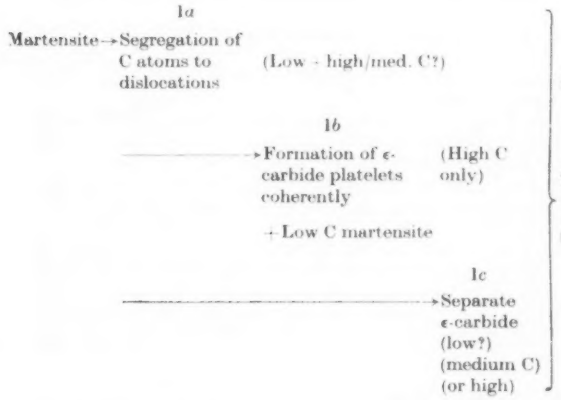
- **'The tempering of 2¼%Cr–1%Mo steel after quenching and normalizing'**, by *R. G. Baker* and *J. Nutting* (1959, **192,** 257–268)
- **'A resistometric study of the first stage of tempering in plain carbon steels'**, by *H. W. King* and *S. G. Glover* (1959, **193,** 123–132)
- **'The influence of tempering-time on some of the mechanical and physical properties of steel'**, by *J. L. Aston* (1959, **192,** 377–382).

Dr **J. Nutting** (University of Cambridge) presented the first paper, Dr **S. G. Glover** (University of Birmingham) the second, and Dr **J. L. Aston** (Head Wrightson and Co. Ltd) the third.

Dr **K. W. Andrews** (The United Steel Companies Ltd): There are a few generalities about these papers which I think it might be useful to make before referring to specific details. If we regard the fully solution-treated austenite or ferrite matrix or a fully quenched completely martensitic structure as representing the base line from which tempering processes or carbide precipitations begin, then the papers in this group form a logical sequence if we consider the paper dealing with the first stage of tempering (King and Glover), next the paper considering the influence of tempering time on properties (Aston), and thirdly the paper by Baker and Nutting which deals in considerably more detail with one particular but in many respects a representative alloy steel. The papers presented in session D (carbides and nitrides) could be regarded as providing information which might be applied at various

stages along the tempering sequence and are complementary.

I do not intend to repeat what the authors have already said, but in the case of the paper by King and Glover there are one or two points I wish to underline. First, they have provided evidence from their electrical resistance measurements of the formation of ε-carbide giving coherency strains during the early stages of tempering *in high-carbon steels* but not in steels with less than, say, 1%C. Second, I believe they have established that *with lower carbon contents* there is another change, associated apparently with the provisional stability of a so-called cubic martensite containing *less than 0·25%C*, whose behaviour is essentially similar to that of ferrite in strain ageing. Furthermore they provide strong evidence that the tempering of low-carbon martensite involves segregation of carbon atoms at dislocations, but not (at least in these early stages) a precipitation. I am not clear as to precisely what they consider happens between 0·25 and 1·0%C but on p.130 they seem to suggest that ε-carbide can form, or the segregation effect can occur as an alternative if more dislocation sites than usual are provided. Presumably at some point with the same steel in this region, segregations of carbon atoms in dislocations could become nuclei of carbide. Have we a tempering scheme like this?



In fact has the first stage of tempering really got three stages 1*a*, 1*b*, 1c? Special circumstances are needed for 1*b*, but otherwise at very short times, and *even at high carbon contents*, could not 1*a* precede the other stages? Again does ε-carbide form for a very short time in the low-C steels? Alternatively is there not very little difference between very thin platelets of ε and similar platelets of cementite? This raises the point whether the changes in direction in King and Glover's Fig.3 for 100° and 120°C could be due to the transition from stage 1*a* to 1c or from 1*a* to 3, or from 1c to 3.

Turning now to Dr Aston's paper, it would be interesting to know if the author has used these results to give values of activation energies and if so, how they have linked up with the relationship of King and Glover. Presumably at the early stages they *should* give similar values, and hence resistance and hardness could be equivalent methods for measuring the progress of tempering. Would the parallelism continue however into the third stage of tempering? The author states that he has used the Hollomon–Jaffe parameter T(20+log t) and found that some modification was required, in particular an additional term. It would nevertheless have been interesting to have seen precisely what these relationships were. I have attempted to show elsewhere[1] that the use of this parameter is not really properly consistent with the usual ideas of rate processes and kinetics: if the parameter *is* obeyed then the simple activation energy treatment does *not* apply, and vice-versa. Incidentally Fig.5 of the paper by King and Glover is actually reminiscent of the time/temperature parameter treatment and in effect uses a simple parameter of the form (AT+log t) where A = constant. This could also be applied to Dr Aston's work. In fact it might be worth investigating whether this treatment fitted many of the numerous sets of results which have been presented in terms of the other parameter.

The third paper by Baker and Nutting also deals with the time/temperature relationship as in Figs.2 and 13, but is mainly concerned with the identity and morphology of the phases. They indicate the basic difference between the two conditions before tempering, namely the quenched or normalized conditions. This accounts for the differences in hardness and carbide formation. Some time ago we reported an X-ray investigation[2] into a steel of this type although the authors do not appear to have seen this. Figure 9 of that paper shows the extent to which the results agree with Baker and Nutting's Fig.2 for the quenched condition. There are some slight differences which are worth mentioning. We do not find ε-carbide but, of course, concede that electron diffraction should be more successful here. Fe_3C appears to persist to higher temperatures and longer times, and Mo_2C appears slightly sooner, in the steel of Baker and Nutting. In our case, however, we appear to have reached a condition at 700°C where only M_6C is left. This seems to us to be the end-point of a good many carbide sequences such as those described by the authors on p.266, which I think fully agree with the one referred to in a paper in the discussion on carbides and nitrides.

Our results for the normalized condition are also closely similar to those of Baker and Nutting, apart from the ε-carbide and Mo_2C 'fringes', and the fact that we seem to have reached the end in 1 000 h at 700°C and M_6C has appeared considerably sooner. The differences might be due to the apparently lower sensitivity of the X-ray methods, but we have often been surprised to find how sensitive they can be. It is more likely, however, that another factor must be considered. The properties of steels of this type, such as $2\frac{1}{4}$%Cr–1%Mo, are known to vary a good deal with the temperature from which they are normalized or quenched. The authors do not give any details of this treatment, and it may well be that the results are considerably affected especially as regards their relative position along the time scale of the various phase regions. An example I have already mentioned is that we have found M_6C appear much earlier in the normalized condition than in the quenched (as above). In some cases undissolved carbide could be present and might affect the interpretation, or the carbide may be dissolved but the matrix has not been by any means homogenized. This raises a problem about this kind of research in general. The heat-treatments likely to give consistent fundamental knowledge are probably not suitable for industrial use. The commercially practicable heat-treatments are not so likely to give consistent results. This dilemma is well known, as the discussion on carbides and nitrides showed.

Lastly I should like to say that the X-ray fluorescent analysis method used by these authors certainly contributed to the value of the work and carried the interpretation further. The quantitative diagrams must, however, be regarded as imprecise, as the authors rightly point out.

Dr **J. Glen** (Colvilles Ltd): Everyone must agree that the paper by Baker and Nutting presents a fascinating picture of the structures which can be obtained on tempering steel; but while not depreciating the importance of this work, I feel that I must take them to task when they state that the simple presence of Mo_2C needles confers high-temperature strength on the type of steel with which they are dealing. This is a much too simplified picture of the complex phenomenon of creep strength.

They also state in the first paragraph of their paper that the term 'stress relieving' is a misnomer. As a practical metallurgist I only wish that they were right. Stress relieving is a treatment carried out to reduce residual stresses and thus the temperature necessary for this treatment may be higher than that necessary to give optimum creep or rupture strengths. Thus it is preferable to use the term 'stress relieving' even though this process is at the same time a tempering treatment.

Since in most practical applications it is necessary to temper

TABLE 1 Residues from 12%Cr steels with varying Ni, Mo, and Co

Steel no.	Method	Cr	Mo	Co	Ni	Fe
1	X-ray (a)	13·3	28·8	5·2	0·9	45·0
	(b)	12·9	31·0	5·9	0·5	42·7
	Chemical	13·4	30·4	5·6	0·3	46·1
2	X-ray	13·9	24·7	6·0	0·6	43·7
	Chemical	14·4	24·0	6·1	0·9	48·0
3	X-ray	13·2	29·9	8·9	0·3	38·5
	Chemical	13·4	29·0	9·0	tr	44·7
4	X-ray	15·9	23·6	3·4	1·6	52·2
	Chemical	15·3	21·0	2·9	1·5	53·6

the 2¼%Cr–1%Mo steel to a temperature above about 650°C, much of the information given in Figs.2 and 13 of the paper is not of great practical importance. What is important is that this steel in the condition in which it goes into service has been tempered to such a degree that the carbide has been spheroidized and the Mo_2C is well defined and in a massive form. With a simpler steel such as ordinary 0·5%Mo steel, the best rupture strength is obtained in the normalized condition and tempering may reduce the rupture strength. It is well known, however, that the short-time creep resistance, which I would prefer to call the initial creep resistance, reaches a maximum after tempering for a few hours at 650°C, and it has been shown by Murray that this maximum corresponds exactly to the optimum coherent hardening due to the presence of Mo_2C. Very prolonged tempering to make the Mo_2C visible reduces the creep strength and also the rupture strength. Since it has already been shown that the 2¼%Cr–1%Mo steel is overtempered by the stress-relieving treatment to give massive Mo_2C carbide, it is to be expected that the initial creep resistance of this steel will be poor, and this is found to be the case.

If the deformation of this steel in practice were to be limited to, say, 0·1% the steel would be a poor one. However, the rupture strength of this steel is very good, indicating that the creep rate continues to fall over very long periods of time. In other words, the latent creep resistance of the steel is high. It must be concluded, therefore, that some structural changes occur during actual testing, after we have heat-treated the steel and examined it under the microscope, and it is these changes which are of essential importance to this steel. Although such changes are not the subject of this paper, I think that until we understand them we shall not know what causes the good rupture strength of this steel.

The paper by Aston is also very interesting since he is attempting to deal with the complexities of his problem. Most of the physicists that I meet try to reduce metallurgy to one simple formula, but the work of Baker and Nutting, among others, shows that as steel is tempered all sorts of changes occur right up the temperature scale. Aston has shown that the rupture strength does not obey a simple rule. Has he tried a similar steel to that used by Irvine and his colleagues, namely a martensitic steel which is also age hardening? With such a steel, would he not find that even his hardness curves would cease to obey the rule which he has laid down?

Dr **H. Hughes** (The United Steel Companies Ltd): I wish to limit my comments to methods of fluorescent analysis. In my experience in this field we have already found that only careful standardization of sample preparation and counting conditions gives results which are directly comparable. We have devised a method[3] on these lines which may be applied to the analysis of residues extracted electrolytically from steels. In its original form 0·1 g of residue was used, but there is no reason why the amount should not be reduced to 25–50 mg without undue sacrifice of accuracy.

TABLE 2 Residue ($M_{23}C_6$) from a 12%Cr–2½%Mo–6%Co steel

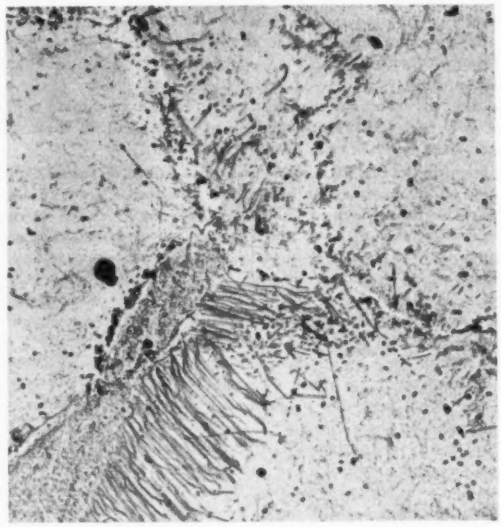
Method	Cr	Mo	Co	V	Fe
X-ray	37·8	21·0	1·2	1	18·5
Chemical	38·5	22·0	1·7	0·9	20·1

U 2¼%Cr–1%Mo *steel cooled at 200°C/h* ×10000

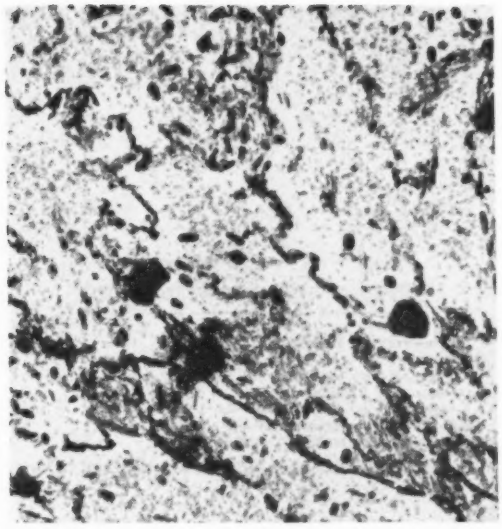
In Table 1 the results of chemical analysis are compared with those of the X-ray method and the comparison is quite good. It all depends on what it is desired to achieve. We are within 5% or so of the actual amount present. Table 2 also illustrates this point. Here we have a residue of $M_{23}C_6$. The apparent level of Mo in Baker and Nutting's paper was somewhat lower than that shown here. I think that this method might be applied in the present case and Figs.8–10 and 20–22 repeated, giving absolute results. It may be argued that at low tempering temperatures fine particles could be lost, but most of the area covered by the figures could be done with confidence.

Written contributions

Mr **F. B. Pickering** (The United Steel Companies Ltd) wrote: I should like to congratulate Drs Baker and Nutting on their



V 0·1%C–½%Mo–1%V *steel tempered for 1 h at 650°C* ×10000



W 1%Cr–1%Mo–0·6%V *steel cooled at 200°C/h* ×10000

very excellent paper in which they have considerably clarified the tempering reactions which occur in this important low-alloy creep-resisting steel. I should like to deal mainly with the effect of rate of cooling on the initial structure of this steel, and also to consider creep-resisting ferritic steels containing vanadium.

A steel of the 2¼%Cr–1%Mo type can exhibit a large number of metallurgical structures when cooled at varying rates, extending from martensite through bainite to polygonal ferrite and pearlite. At the slower rates of cooling, fringes of Mo_2C carbides can form either at the bainite/ferrite interface or the pearlite/ferrite interface as shown in Fig.*U*. Also, considerable quantities of Mo_2C carbide can precipitate within the ferrite.

A similar situation often occurs with vanadium steels. The formation of fringes of V_4C^3 carbides is shown in Fig.*V* for a Mo–V steel which had been normalized and then tempered at 650°C. These fringes occur together with V_4C_3 precipitates within the ferrite, and the structure shows that they probably occurred on dislocations. During cooling at 200°C/h, steels containing 1%Cr–1%Mo–0·6%V show bainitic structures which are heavily tempered to produce many small particles of V_4C_3 (Fig.*W*). At a slower cooling rate of 60°C/h, there may be many fringe carbides (Fig.*X*(*i*)) and also V_4C_3 precipitates within the ferrite (Fig.*X*(*ii*)).

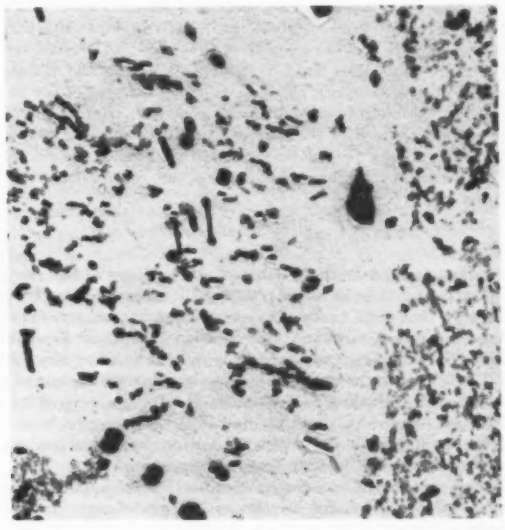
These effects emphasize the fact that at slow cooling rates, which are often employed in commercially treated power-plant materials, very considerable variations can occur not only in the transformed structure, but also owing to the effect of tempering produced by the slow cooling rate. The subsequent behaviour of such structures both during testing and service can be very different because their subsequent hardness changes during tempering are also different. It is essential therefore that the utmost care is taken when comparing data on such steels to make sure that the initial structures were the same.

Gemmill and Murray[4] have shown some effects of prior structure in 2¼%Cr–1%Mo steels on the 1000-h rupture strength at 565°C. In the normalized and tempered condition the strength was about 12 tons/in² and the structure completely tempered bainite, but after furnace cooling at 33°C/h the strength was only 10 tons/in², and with an even slower rate of cooling of 15°C/h had decreased further to about 8 tons/in². After tempering each of these structures for 100 h at 700°C to simulate long-time exposures at 565°C, the rupture strengths of each structure were identical at about 7 tons/in². The point of this observation is that the presence of polygonal ferrite and precipitation in the initial condition can materially weaken the short-time strength, but the ultimate long-time strength will be largely unaffected by the initial structure.

Finally, we have evidence to show that normalized material consisting of 100% bainite tempers rather more slowly in the temperature range 550–650°C than does water-quenched, i.e. martensitic, material. This may well mean that normalized bainitic structures have better strengths than martensitic structures, at high temperatures. We have not yet examined the tempering reaction of these steels using the electron microscope but I would value the authors' opinions on the apparently different tempering characteristics of fully bainitic structures compared with fully martensitic structures.

Dr **Mats Hillert** (Swedish Institute for Metal Research, Stockholm, Sweden) wrote: Although the main features of the first stage of tempering are fairly well established there are still problems which need further experimental work in order to be resolved. King and Glover discuss two of these in greater detail in their introduction and later, from their own

X (i) 1%Cr–1%Mo–0·6%V *steel cooled at 60°C/h* ×10000



X (ii) *As Fig.X (i)* ×10000

TABLE 3 Activation energy, kcal/mole, for four steels

%P	0·19%C	0·47%C	0·69%C	0·92%C
0·0625	4·5			
0·125	6	9		
0·25	8	10		16
0·5	11	13·5	15·5	17
1·0	17	17·5	17·5	19
2·0		20·5	19·5	21·5
4·0		23	21·5	22·5
6·0		23·5	20·5	24
10·0		22·5	22·5	25
K+G	19·45	22·24	23·66	25·51

experimental results, they arrive at certain conclusions. However, it may seem that the same experimental results could as well be used as support for the opposite view.

1. The first problem concerns the precise value and significance of the activation energy for the first stage of tempering. Widely conflicting values have been reported, many of them surprisingly high in comparison with the activation energy for diffusion of carbon in ferrite. Lement and Cohen[5] have suggested an explanation of the high values based on the assumption that the value of the activation energy for diffusion of carbon in martensite is dependent on the carbon content of the martensite. It should thus become identical to the value for ferrite in that limit of vanishing carbon content only. This concentration dependency certainly needed to be tested experimentally and, moreover, as King and Glover themselves state: 'It is not at all certain that the assignment of a unique value to the activation energy was justified'.

In evaluating their data, King and Glover disregard the first part of their experimental curves and arrive at the conclusion that for each steel there is a unique value of the activation energy for the first stage of tempering. They find that for a series of steels the activation energy varies with the carbon content of the steel, c_0, according to an equation which we can write

$$Q_{c_0}^{\text{1st stage}} = 8500c_0 + Q_{\text{ferrite}}^{\text{diffusion}}.$$

They explain this variation in terms of the tetragonality of martensite, which is known to increase with the carbon content. The assumption is that diffusion in martensite can be described by an equation

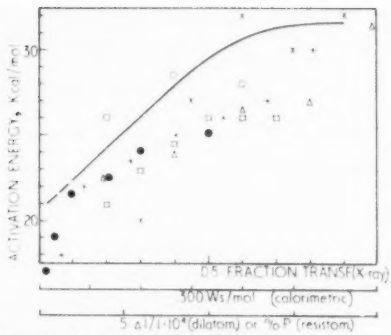
$$Q_c^{\text{diffusion}} = \mathrm{K}c + Q_{\text{ferrite}}^{\text{diffusion}}.$$

This model was studied by myself in a recent paper[6] and the following results were obtained. (*a*) A theoretical evaluation of the constant K yielded a value of $\mathrm{K} = 1700$. (*b*) The diffusion coefficient may thus vary considerably along the diffusion paths from regions with the initial carbon content of the steel, c_0, up to the neighbourhood of the ε-carbide particles, where the carbon content of the martensite might be quite low. The activation energy for the first stage of tempering should then, if controlled by diffusion of carbon, be represented by an expression

$$Q_{c_0}^{\text{1st stage}} = \mathrm{K}.c_{\mathrm{m}} + Q_{\text{ferrite}}^{\text{diffusion}},$$

where c_{m} represents the 'effective' mean value of the carbon content along the diffusion paths. (*c*) The ratio c_{m}/c_0 is considerably less than unity. On comparing with the results of King and Glover, one thus obtains $\mathrm{K}c_{\mathrm{m}} = 8500.c_0$ or $\mathrm{K} > 8500$, which is in sharp contrast to the theoretical value of $\mathrm{K} = 1700$. (*d*) The first stage of tempering is too rapid to be explained by an ordinary diffusion process with $\mathrm{D} = \mathrm{D}_{\text{ferrite}} \cdot \exp -8500c_0/\mathrm{RT}$. With regard to these results, it is tempting to look for another explanation of the high activation energy values found experimentally for the first stage of tempering.

In view of recent results by Gerdien,[7] some explanation is needed of the variation of the activation energy with the degree of transformation of a certain steel as well as with the carbon content in a series of steels. Further support for this



○ dilatometry by Roberts[8] on 0·78%C
+ resistometry by Jellinghaus[9] on 0·79%C
— dilatometry by Gerdien[3] on 0·79%C
● resistometry by King and Glover on 0·92%C
□ X-ray study by Kurdjumov[10] on 1·0%C
× calorimetry by Krisement[11] on 1·03%C
△ dilatometry by Roberts[8] on 1·43%C

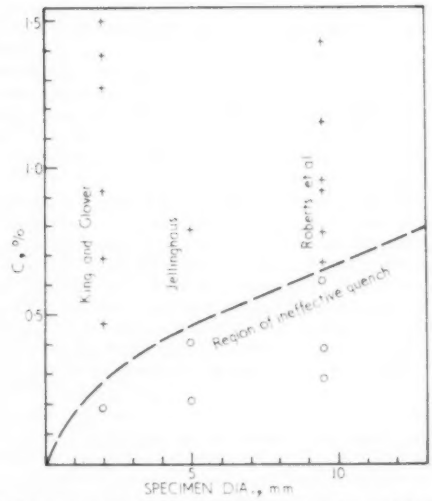
Y *Comparison of data from different experimental techniques and steels*

suggestion can be obtained from the experimental data presented by King and Glover. If the data in their Table II are used for evaluating an individual activation energy value for each degree of transformation, a considerable variation will be found, as demonstrated in Table 3. Table 3 also gives the values evaluated by King and Glover and by them considered to be unique values characteristic of the whole first stage of tempering in each steel.

The very rapid increase up to about 1·0%P may perhaps be explained by a temperature-dependent error in the measurement of the initial resistivity of the specimens. However, it is difficult to escape the impression that there is a significant increase of the activation energy as the reaction proceeds. In fact, if the experimental data published by other authors are re-evaluated in the same way, they all seem to indicate a variation of about the same order of magnitude. This is demonstrated by Fig. *Y*, where data from different experimental techniques (calorimetric, dilatometric, resistometric, and X-ray intensity) are compared.

If one accepts this strong increase of the activation energy as the reaction proceeds, one might also expect that the activation energy evaluated at the end of the first stage of tempering, should increase with the carbon content of the steel, because the reaction proceeds farther the more carbon there is to be precipitated. The variation found by King and Glover may thus be explained once we have found the explanation for the variation with the degree of transformation. It seems that attention should at present be focused on the latter phenomenon.

2. The second problem concerns the carbon content of the low-carbon martensite which forms together with the ε-carbide. This martensite has been found to have a tetragonality corresponding to a carbon content of about 0·25%. It has been suggested that this represents the carbon content of martensite in metastable equilibrium with ε-carbide. The total change in length, resistivity, or heat content during the first stage of tempering of a steel with c_0% carbon should thus be approximately proportional to $c_0 - 0{\cdot}25$. In particular, a steel with less than 0·25%C would not show the ordinary first stage of tempering at all. This suggestion has now been tested experimentally by King and Glover using a steel with 0·19%C and from the small change in resistivity of this steel, only 1·8%, they conclude that the first stage of tempering has not operated here. However, in spite of this, it may be worthwhile testing the other possibilities, e.g. that the total change during the first stage may be proportional to the total carbon content c_0. If the total change in resistivity for the different steels examined by King and Glover is plotted against the carbon content of the steels, it will be found that all the points, except the one for 0·19%C, fall very close to a straight line through



Circles represent values considerably below the straight lines

Z *Relation between carbon content and diameter of specimen*

the origin. For 0·19%C the straight line predicts a change in resistivity of 5·6% but the experimental point lies at a third of this value, 1·8%. This reduction by two-thirds does not seem large enough to warrant the conclusion that the ordinary first stage of tempering is completely absent in this steel. A much more attractive explanation would be that most of the martensite in this steel is formed at such high temperatures during the quench that it is considerably tempered before the steel has reached room temperature. Judging from the amount of tempered (i.e. dark) martensite which can be seen microscopically after etching a polished section of a quenched sample of such a steel, it seems quite reasonable to assume that only a third of the martensite is formed at sufficiently low temperatures to remain untempered during the quench. Only this one third of a specimen would later undergo the first stage of tempering, giving a third of the resistivity change expected for the whole specimen. If this explanation is correct, the rate of cooling during the quench should be very important. In fact, this seems to be borne out by experiments. If the results obtained by Roberts *et al.*[9] and by Jellinghaus[9] are also represented by straight lines through the origin, it is again found that the proportionality holds only above a certain carbon content. This critical carbon content seems to depend on the diameter of the specimens used. In Fig. Z the circles represent experiments which have yielded values considerably below the straight lines. These circles fall inside a region which looks like a region of ineffective quench. The shape of this region demonstrates that one must choose thinner specimens the lower the carbon content is, if one wants to avoid tempering during the quench. This would be expected owing to the increase of the M_s temperature with decreasing carbon content.

If one accepts that the total change during the first stage is proportional to the carbon content, c_0, one cannot accept that low-carbon martensite with 0·25%C is always formed together with the ε-carbide on tempering martensite. The question then arises how the X-ray results concerning the tetragonality of the low-carbon martensite should be interpreted. There seem to be two possibilities. (*a*) The X-ray data may not be accurate enough to exclude the possibility of a variation of the carbon content of low-carbon martensite with the carbon content of the steel. Information is available from only one steel of low carbon content (0·32%C + 19%Ni) and a low tetragonality corresponding to 0·10–0·20%C was here reported.[12] (*b*) The tetragonality found by the X-ray method may be caused partly by the coherency between the ε-carbide and the matrix lattice. The best method to calculate the carbon content from the X-ray data would therefore be based on the volume of the unit cell $a^2.c$ and not on the tetragonality c/a. Table 4 shows the results of different calculations based on the data by Kurdjumov and Lyssak[13] on a steel with 1·4%C and by Werner[14] on a steel with 1·43%C.

TABLE 4 Results of calculations based on data from different workers

Temp., °C	Time, time, h	c	a	Carbon content calculated from c/a	c	a^2c
Kurdjumov *et al.*[11] (1·4%C)		kX	kX			
125	1	2·886	2·848	0·29	0·22	−0·01
150	1	2·886	2·852	0·27	0·22	0·08
175	1	2·884	2·857	0·21	0·20	0·17
Werner *et al.*[12] (1·43%C)		Å	Å			
65	1 200	2·895	2·855	0·31	0·25	0·08
93	80	2·892	2·856	0·29	0·22	0·07
121	4	2·896	2·853	0·33	0·26	0·04
149	1·5	2·895	2·855	0·31	0·25	0·08
204	0·0056	2·893	2·851	0·33	0·23	−0·03

The agreement between the two sets of data is quite good. The values in the last column indicate that the carbon content of the so-called 'low-carbon martensite' may be much lower than previously assumed. It seems that the main part of the tetragonality is caused by some other factor, e.g. coherency. As a consequence, one should expect to find the ordinary first stage of tempering in low-carbon steels as well as in steels of higher carbon content. In this connexion, the observation of ε-carbide in a tempered 0·18%C steel by Modin[15] may be worth noting.

Authors' replies

Dr **R. G. Baker** wrote: We are very pleased to hear from Dr Andrews that his work agrees fairly closely with our own. Our evidence for the presence of ε-carbide is largely circumstantial and in fact we were unable to get satisfactory X-ray or electron diffraction evidence of the structure of the carbide which was present. However, we feel that there must be, if not ε-carbide, then some other intermediate carbide which is changing to cementite on further tempering. We should like to thank Dr Andrews for drawing our attention to his work.

The heat-treatments used are given in Appendix I to the paper, p.267. We cooled at 300°C/h from the heat-treatment temperature of 960°C to get a normalized condition, since the specification for this type of steel is that it shall be cooled between 200° and 400°C/h.

In reply to Dr Glen, we felt 'stress relieving' to be a misnomer not because we did not believe that stresses were relieved but because we thought that the predominant changes were due to tempering, so that the term 'stress relieving' was rather misleading. We believe that the optimum tempering treatment on the normalized steel is not necessarily that which gives the highest initial high-temperature strength but that which gives the maximum strength consistent with thermodynamic stability during creep. This is the essential difference in behaviour between the quenched and tempered and the normalized and tempered steel; the latter has much greater thermodynamic stability for a given tempering treatment although its initial strength may be lower. We cannot see how the steels of this type, in which most of the precipitation of molybdenum carbide appears to have occurred, could have much latent creep resistance left.

In reply to Dr Hughes, we used the technique of fluorescent analysis on the extraction replicas, realizing that it had certain drawbacks but feeling that it was better than the alternative method of electrolytic extraction and fluorescent analysis of the extracts. It is possible to calibrate the method which we used and we have now done this for a limited range of compositions, but unfortunately not in time to include the results in the paper. The results are merely expressed as a percentage of the counts received without any correction for absorption and the various other errors present. If the corrections are made, the figures for chromium are substantially raised and we

believe they will be comparable with those obtained by Dr Hughes.

We should like to thank Mr Pickering for his kind remarks and for his most interesting written discussion. He notes that fringes of Mo_2C and V_4C_3 are likely to form at the slower rates of cooling and we believe this observation to be consistent with the mechanism we proposed to explain the formation of Mo_2C. In an earlier paper[16] we reported V_4C_3 fringes which had formed in a steel containing vanadium after air cooling from 1150°C and then tempering at 600°C.

We confirm Mr Pickering's remarks concerning the greater suitability of tempered bainitic structures for use at high temperatures. From the results of this paper it seems likely that fully bainitic structures should have a greater thermodynamical stability than the tempered martensites. Thus they should have better long-time creep characteristics.

Dr **H. W. King** and Dr **S. G. Glover** wrote: We wish to thank Dr Andrews and Dr Hillert for their valued comments. The generalized scheme of tempering is, we believe, as depicted by Dr Andrews. It is difficult, however, to determine the complete sequence under a particular set of conditions for usually one of the steps predominates and others, if they occur, may do so either too slowly or too rapidly to permit detection. Oversimplification is the almost inevitable consequence. Thus stage 1*a* may precede the others under all conditions of carbon content and tempering temperatures but we do not know this. Also it is probably true that ϵ-carbide and cementite are false distinctions when applied to precipitates in thin plate form; the somewhat puzzling conflict of recent evidence on carbides in aged ferrite is perhaps testimony of this. Nevertheless the changes in direction of the 100° and 120°C tempering curve for our 0·69%C steel noted by Dr Andrews arise, according to the most reasonable assumptions based on earlier diffraction data, from the transition 1*c* to 3.

Dr Hillert has pointed out, from an examination of the somewhat incomplete data listed in Table II of our paper, that the activation energy of the first-stage process is initially rather low and increases rapidly as tempering proceeds. However, a study of the complete data reveals, as demonstrated in Fig.7 of our paper, that after the initial increase the activation energy approaches a constant value which is maintained until the onset of the third stage of tempering. It is this 'steady-state' activation energy which we have evaluated for the seven plain carbon steels. It is of interest to note that in spite of the relatively large scatter, the data of other investigators quoted by Dr Hillert in his Fig.*Y* also show this trend in approaching some constant value as the reaction proceeds, in particular the precise data of Kurdjumov and Lyssak. Further, these 'steady-state' values of activation energy agree closely with our data when plotted against the carbon content of the martensite, as shown in our Fig.8. Since the reason for the low initial activation energy of the process and its subsequent increase to some steady value appears to be of considerable interest, we shall deal with it in detail.

Some years ago Kurdjumov and Lyssak[13, 10] suggested that these peculiar characteristics of the first stage of tempering may be understood by considering a mechanism based on the Thomson-Freundlich formula:

$$\ln\frac{s}{s_\infty} = \frac{2\mu\tau}{RT\rho r} \qquad \ldots\ldots(1)$$

in which r is the radius of the precipitated particle, s_∞ the saturation concentration of carbon in the primary martensite, s the saturation concentration of carbon in the precipitate, and μ, ρ, and τ its molecular weight, density, and surface tension respectively. According to their mechanism, the kinetics of the first-stage process should be controlled by the formation of nuclei of ϵ-carbide, the growth of which is restricted by the formation around them of a narrow zone where the carbon content of the original martensite has been depleted to a limited extent. The carbon content of the bulk of the martensite matrix would thus remain unchanged, in accordance with observations mentioned above. Since diffusion was discounted in their mechanism, the rapid growth of a particle would be possible only so long as the martensite adjacent to the ϵ-carbide nucleus is saturated with carbon relative to the precipitate; i.e. the carbon content exceeded s in equation (1). Although they originally suggested,[13] on the basis of the earlier work of Kurdjumov and Sachs,[17] that this mechanism applied to the entire first stage, their later experimental results[10] show that the predicted reaction-kinetics apply only to the initial moments of tempering.

The mechanism of Kurdjumov and Lyssak has tended to be forgotten since Roberts *et al.*[8] and Lement and Cohen[5] have introduced mechanisms for the first-stage process based entirely on the diffusion of carbon through the martensite. Owing to the particular method used by Roberts *et al.* to study tempering, they were unable to follow the very early stages; hence, it is not surprising that these authors dismissed nucleation as a controlling mechanism. Nevertheless, the evidence from electron microscopy[18] shows that the first stage of tempering takes place by the limited growth of a large number of nuclei of ϵ-carbide, in contrast to the extended growth of relatively few nuclei in other precipitation processes. Hence, in its initial moments at least, the first-stage process must be controlled more by the number of nuclei of ϵ-carbide forming in unit time, per unit volume, according to the conditions of equation (1), than by the diffusion of carbon through the martensite. Thus we suggest that the low initial activation energy of the process is to be associated with the formation of nuclei of ϵ-carbide which, as discussed in our paper, maintain lattice coherency with the martensite matrix. Once the carbon content of the martensite in the zones surrounding the ϵ-carbide particles is depleted to below the value of s in (1), it will become essential for carbon to diffuse through the bulk of the martensite in order to maintain the reaction. Hence, the activation energy will be expected to increase to some steady value which may be associated with the diffusion of carbon in martensite. Since the carbon content of the bulk of the martensite will be unaffected by the initial localized precipitation of ϵ-carbide, it is in order to associate the steady-state activation energy with the diffusion of carbon through the primary martensite.

Dr Hillert has compared our expression for the dependence of the activation energy for carbon diffusion in martensite (Q_c) on carbon content with that which he has derived theoretically. The latter is based on Zener's calculation of the energy difference, U, between two neighbouring interstitial sites in martensite of carbon content c, due to the tetragonality of the lattice. This is given as $NU = 6700c$ cal/mole, N being Avogadro's number. In arriving at the increase in Q_c due to the tetragonality, Hillert assumes that the increase in the height of the energy barrier between the neighbouring lattice sites is $\frac{1}{2}U$. Thus the quantity to be compared with the 8500c of equation (3) of the paper is $NU/2 = 3400c$ rather than 1700c which is the latter divided by the gas constant and is the form in which the constant correctly enters the equation for the diffusion *coefficient* of carbon in martensite in Hillert's original presentation.[6] The remaining discrepancy of a factor of 2 or 3 does not, we consider, offer a serious challenge to our experimental values for Q_c in view of the uncertainties of the theory. Finally, on this subject, we should like to emphasize that, for reasons given in the last paragraphs of the paper, we do not associate the term 18000 cal/mole in equation (3) with the activation energy for carbon diffusion in ferrite as is tacitly ascribed to us.

The second point raised by Dr Hillert in his careful appraisal concerns the tempering of low-carbon martensites. We agree with him that some tempering must have taken place in the 0·19%C steel during the quench and admire his ingenious correlation of the degree of such tempering, carbon content, and specimen size presented in Fig.*Z*. However, it seems doubtful whether this would remove two-thirds of the original carbon from solid solution and even if this amount were precipitated during the quench, it does not follow deductively that the remainder will undergo first-stage tempering. Our proposal that the martensite in this steel undergoes a form of

strain-ageing was based not only upon the resistivity change being anomalously feeble but also on the observed initial time exponent of 2/3.

Dr Hillert's alternative interpretation of the diffraction results points strongly to the so-called 'low-carbon martensite' having a carbon content considerably below the previously accepted 0·25%. This, together with the detection of ε-carbide in a tempered 0·18%C martensite, would remove the main obstacles to ascribing the resistivity decay in our 0·19%C steel to normal first-stage tempering. The question is whether this would be correct, for the points made by Dr Hillert do not so much invalidate our proposal as suggest that it is unnecessary to make a distinction between the tempering mechanism in steels according to whether the carbon lies above or below 0·25%. Experimental evidence is lacking but it might be speculated that, as proposed for ferrite,[19] carbon segregates to dislocations and, when the initial carbon content is low, only after the matrix has already been largely drained of carbon does ε-carbide form.

Dr **J. L. Aston:** I should like to thank those who have contributed to the discussion for their helpful remarks.

In reply to Dr Glen I have not carried out tests of this type on any steels except those for which results are quoted.

As a result of the suggestion by Dr Andrews I have used the hardness results to plot log t against the reciprocal of the absolute temperature. The results for steel *A* (3%Ni) are not suitably distributed for this, but do not suggest a series of parallel lines. Steel *B* (0·4%C–1·4%Mn) has a better distribution of data, and the points are sufficiently near to representing a series of parallel lines for the value of the activation coefficient to be determined as 72×10^3 cal/mole. Steels *C* (0·47%C) and *D* (0·46%C–1·00%Cr) are well represented by parallel straight lines, especially the latter. The activation energies are 68×10^3 and 77×10^3 cal/mole respectively. These values are clearly different from those obtained by King and Glover, but they relate to much fuller tempering.

Dr Andrews also asks about the empirical relationships which were determined for properties in terms of tempering temperature and time. These were taken out of the paper in the interests of brevity, but I quote three here, as examples. Time is in seconds, and temperature in °C.

Steel C (0·47%C)

Interpolation range: 5 s–1 day 400–700°C

Hardness (DHN) =

$326 - 32(\log_{10}t - 2{\cdot}8) - 0{\cdot}75(\theta - 550) + 0{\cdot}07(\log_{10}t - 2{\cdot}8).(\theta - 550)$

The coefficient of the term in $\log_{10}t.\theta$ is significant at the 5% level but not at the 1% level. The others are significant at the 0·1% level.

Residual standard deviation = 17 DHN

Reduction in area (%) = $48{\cdot}5 + 3{\cdot}7(\log_{10}t - 2{\cdot}8) + 0{\cdot}07.(\theta - 550) - 0{\cdot}01(\log_{10}t - 2{\cdot}8).(\theta - 550) - 1{\cdot}1\ (\log_{10}t - 2{\cdot}8)^2$

The coefficient of the term in $\log_{10}t.\theta$ is just significant at the 5% level. The coefficient of the term in $(\log_{10}t)^2$ is significant at the 5% level, but not at the 1% level. The others are significant at the 0·1% level.

Residual standard deviation = 3·1%.

Steel D (0·46%C, 1·00%Cr)

Interpolation range: 4s–6 days 400–700°C

Hardness (DHN) = $386 - 29{\cdot}6(\log_{10}t - 3{\cdot}15) - 0{\cdot}75\ (\theta - 550) - 0{\cdot}05(\log_{10}t - 3{\cdot}15).(\theta - 550)$

The coefficient of the term in $\log_{10}t$. θ is significant at the 0·1% level, as are the others.

Residual standard deviation = 7·6 DHN.

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Correspondence on the paper

Abraham Darby and the beginnings of the coke-iron industry*

Professor **R. A. Mott** (University of Sheffield) wrote: When writing this interesting paper Dr H. R. Schubert* was unable to see my paper 'Coalbrookdale: 1709'.[1] In this I dealt with some of the doubts which Dr Schubert expressed, e.g. the statements of Abiah Darby and Samuel Smiles. He will also, no doubt, now have seen that what he terms the 'serious objections' of Flinn[2] have been dealt with in my rejoinder,[3] where they are shown to be both trivial and misleading.

Dr Schubert suggests that some of the coal supplied to Coalbrookdale in 1709 was used for house-heating, some in the drying and curing of malt, and some for the manufacture of brass. He concludes that 'until more evidence is unearthed, it is impossible to fix a precise date for the first successful application of coke as a fuel for iron-smelting'. This view must be refuted.

In the Coalbrookdale Company's Stock Book for 1718–27[4] there are records for four-weekly periods of the delivery to the old and new blast-furnaces of large coal (which was coked) and lump coal (presumably for roasting the ore), of large (and some lump) coal to the air-furnaces (where pig iron was remelted for smaller castings), and of deliveries under 'general charges' which are identified in other entries as being 'to ye houses' of Mary Darby (two periods only), John Darby, Thomas Baylies, and Richard Ford. In 1719 (O.S.), the first full year of the records, coal 'to ye houses' amounted to only 1·25% of the total deliveries. Since any coal delivered for malt-making or brass-making would be included in 'general charges', it is clear that in 1718–27 the amount of coal supplied other than to the blast-furnaces and air-furnaces was negligible. There is no reason to suppose that such is not the case for 1709.

In fact it cannot be true, as Dr Schubert suggests, that 'coal

* H. R. Schubert: *JISI*, 1959, **193**, Sept., 1–5.

or coke was absorbed in the drying and curing of malt'. Malt was made in England at that time by soaking barley in water, allowing it to ripen, and then curing (or drying) in a coke-fired furnace. At Coalbrookdale in 1708–9, however, malt was purchased, not made, as Dr Schubert himself records. Similarly the brass (used for small patterns) was purchased in 1709 (and later). Moreover, Schubert fails to note that in using a reverberatory furnace for making calamine brass (as Abraham Darby did at Bristol) it was essential to put the raw materials, copper, roasted calamine (ZnO), and charcoal, into closed pots, the process being one of reduction. Schubert describes the furnace as one for the smelting of lead or copper ore; in fact it could not have been used for making brass, for which access above the furnace for the insertion and extraction of the pots was essential.

In 1709 Abraham Darby purchased 'coles' at a cost which accounts for 862 stacks (1510 tons), the units of measurement showing it to have been mineral coal; he made 90 tons of iron, giving a ratio of 16 tons of coal per ton of iron which is similar to that which obtained in 1723, when it cannot be disputed that only coke was used in the blast-furnaces. There are no entries in the accounts[5] for the purchase of cords of wood or for the charges for cutting, cording, 'coling', and transport in 'dozens and baggs' such as occur in the records for 1720–22[4,6] when charcoal was used in the forge. If Abraham Darby had used charcoal his rate of production in 1709 would have been over 10 tons/week instead of the value of 2·1 tons/week which we know he actually obtained. Moreover, his process would have been uneconomic because of the high, and known, price of charcoal and the prices he received for his products. Furthermore, he could not have made successfully the wide range of thin castings which we know he made and which required the higher silicon content of coke-iron compared with that of charcoal-iron.

The patent of 1695 for smelting 'with pitt coales or sea coales charked' is of no significance since the assessment made at that time was whether the patentee had put his patent to use. In any case, Abraham Darby was covered by his own patent.

In conclusion it may be said that the subject demands a book for the critical consideration of the evidence and of the objections, usually trivial, so often raised. That over 250 people were sufficiently interested to attend the meeting at Birmingham suggests that there is wide interest in the subject. Since this English enterprise of 250 years ago had more influence than any other single industrial development, it seems appropriate that it should be adequately documented by such means.

AUTHOR'S REPLY

Dr **H. R. Schubert** wrote in reply: The numerous assumptions Dr Mott makes in his paper read at Birmingham and his constant use of terms such as 'presumably', 'likely', 'probably', 'must be', etc. do not make his conclusions any more convincing than his previous papers. In addition, he still does not refrain from the dangerous method of using accounts for the period after 1718 to illuminate the account of 1709. Such a method sometimes has a rather perplexing effect. The 'wood coller', who worked for 17 days in the second blast of 1709 would normally be taken for a charcoal burner, which, however, would disagree with Mott's assumption that no other fuel but coke was applied for smelting. Thus the collier's activity, according to Mott,[1] was confined to the making of fagotts, i.e. hurdles required as wind shields for coke heaps. This he explained by quoting an entry of 1732 referring to a 'wood coller' who made such faggots at such a date. But this later collier also made charcoal.

In his so-called 'rejoinder'[3] Mott has not succeeded in any way to invalidate Flinn's objections. Starting with the ancient Greeks and Romans he traces superficially the development of the terms used for coal, and arrives at the conclusion that to the people of Shropshire at the end of the 17th century 'coles' must usually have meant mineral coal. This is contradicted by an assignment of 23 Feb. 1719/20 referring to a share in the stocks of Weighton furnace, in Shropshire, close to Coalbrookdale, 'consisting of ready money, metall, coles, cordwoods, and utensils'.[7] Weighton furnace was still operated exclusively with charcoal in 1755 at which date Reinhold Angerstein, as official of the Jernkontoret at Stockholm, visited the works. Mott's argument[8] that 'in the mineral coalfields the unqualified term "cole" meant mineral cole by the end of the 17th century', cannot be maintained. In the South Wales coalfields, e.g. at Llanelly and Pontypool, in 1704 'cole' stands for charcoal while mineral coal is termed 'stone coal'.[9] There are other examples of similar terms. In any case, the term 'cole' cannot be used as Dr Mott uses it, as an argument for an exclusive application of mineral coal coked of charcoal.

It is difficult to understand why he turns the charcoal blast-furnace at Dolgyn near Dolgelly, Merionethshire, into 'an air (reverberatory) furnace to remelt pig iron for castings.'[1] It is in complete disregard of the facts, laid down in Kolsall's Diary. The furnace was not built in 1719, in the spring of which the first campaign producing 446 tons of pig iron began, but building began in the early summer of 1718. There are many references to hearth, tuyère, and boshes (still visible in the ruins). In the subsequent campaigns pig iron was produced exclusively, but not castings.

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NEWS | *Announcements and News of Science and Industry*

THE IRON AND STEEL INSTITUTE

President-Elect

At a meeting of the Council of The Iron and Steel Institute held on 13 July 1960, it was agreed unanimously to propose Sir Charles GOODEVE, O.B.E., D.SC., F.R.S., for nomination as President-Elect of the Institute at the Autumn General Meeting in November. Subject to election at the Annual General Meeting in May 1961 Sir Charles will then succeed Mr W. F. CARTWRIGHT as President and will thus be in office at the time of the Institute's proposed Special Meeting in the USA in October 1961.

Changes on Council

The following changes were announced after the meeting of Council on 13 July:

Vice-Presidents

Mr W. C. BELL (*Member of Council*) managing director iron- and steelworks, Stewarts and Lloyds Ltd.

Professor A. G. QUARRELL (*Member of Council*) professor of metallurgy, University of Sheffield.

Honorary Vice-Presidents

Major W. R. BROWN, D.S.O. (*Member of Council*) director, Ashmore, Benson, Pease and Co. Ltd.

Mr T. JOLLY (*Member of Council*) joint managing director, Guest Keen Iron and Steel Co. Ltd.

Members of Council

Mr M. A. FIENNES (*Member*) managing director, Davy United Ltd.

Mr G. Wilton LEE (*Member*) chairman and managing director, Arthur Lee and Sons Ltd.

Mr P. M. THOMAS (*Member*) managing director, William Beardmore and Co. Ltd.

The Medal of The Iron and Steel Institute

The Council of The Iron and Steel Institute has established a new gold medal, to be known as 'The Medal of The Iron and Steel Institute'. This Medal is to be presented on special occasions to societies which, in the opinion of the Council of the Institute, have rendered exceptional service to science or industry over a lengthy period.

The first Medal of The Iron and Steel Institute was offered to the Royal Society, in recognition of the Society's unique service to science over a period of 300 years.

It was presented by the Hon. R. G. LYTTELTON, Past-President of The Iron and Steel Institute, at the Tercentenary Celebrations of the Royal Society on Monday, 18 July 1960.

Iron and Steel Institute Prize 1960

The Iron and Steel Institute prize of 10 *gns* has been awarded to Mr NEIL J. CAVAGHAN, B.MET. (HONS) of the University of Sheffield.

NEWS OF MEMBERS

Mr **G. H. Davison** has retired from the post of secretary of the research and development department of The United Steel Companies but continues to serve in an advisory capacity.

Mr **R. Fannon** is now at the head office of the General Dynamics Corporation in Chicago.

Mr **A. F. Jessop** has been appointed manager of the Appleby melting shop of the Appleby-Frodingham Steel Company.

Mr **R. Johnson** has been appointed melting shops manager at the Appleby-Frodingham Steel Company.

Mr **T. S. Kilpatrick** has been appointed director and general manager of Steel, Peech and Tozer.

Dr **A. B. Kinzel** has been elected a member of the US National Academy of Sciences.

Mr **J. D. D. Morgan** has been appointed to the research board of the British Welding Research Association.

Mr **W. L. Willsher** has been appointed manager of the Frodingham melting shop of the Appleby-Frodingham Steel Company.

Mr **R. Wogin** has been appointed assistant works manager (steel) at the Appleby-Frodingham Steel Company.

Obituary

Professor **Kurt Henrik Amberg** (*elected* 1954) of Stockholm, in November 1959.

Karl Anders Fredrik Göransson (*elected* 1926) of Sandviken, Sweden, on 13 June 1960.

Dr **Earle Clement Smith** (*elected* 1925) of Cleveland, Ohio, USA, on 20 May 1960.

AFFILIATED LOCAL SOCIETIES

Ebbw Vale Metallurgical Society

At the annual general meeting of the Society on 9 June, R. C. Powell, B.SC. was elected president. Other elections were:

Vice-President F. W. Lewis

Honorary Treasurer L. G. Hobbs

Members of Committee R. M. Bethed, J. B. Davies, M.SC., T. E. Davies, P. T. Edwards, W. G. Edney, A. W. Ellis, I. Giddings, D. J. Jones, B.SC., A. J. Meo, D. L. Mills, J. T. Mills, H. Padget, J. H. Powell, M.C., J. F. Smith, J. B. Thickens, B.SC., W. B. Williams.

Joint secretaries D. W. Evans, B.SC., and W. E. Williams.

Staffordshire Iron and Steel Institute

Officers and Council elected for the year 1960–61 are

President F. V. Wright

Senior Vice-President E. B. Gibbons, J.P.

Junior Vice-President G. T. Hampton

Trustees P. Jump, H. E. Cookson, T.D., W. H. Lewis

Hon. auditors N. S. B. James and L. Chater

Hon. treasurer G. E. Lunt

Members of Council W. M. S. Walker, PH.D., M.SC., W. K. V. Gale, F. A. Meachem, G. B. Forster, B.SC., M. Lloyd, H. J. E. Jones, M.C., B.A., L. Harper, T. H. Morfey, C. Macquarie, B.SC., G. Harrison, J.P.

Members of Council (*past-presidents*) A. Biddulph, K. H. Wright, K. G. Lewis, M.SC., M.SC.(TECH), A. W. Shore, W. H. B. Wesson, T.D., L. W. Law, M.A.

Secretary G. Riley.

Mr W. F. Cartwright, D.L., J.P. (President of The Iron and Steel Institute) receiving the Federico Giolitti Steel Medal from Professor Felice De Carli (President of the Associazione Italiana di Metallurgia) at the Joint Meeting in Milan in May.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Staff appointment

Mr W. E. Duckworth, M.A., A.I.M., has been appointed Head of the Metallurgy (General) Division of BISRA. A graduate of Queens' College, Cambridge, Mr Duckworth has specialized in corrosion and fatigue problems in cables and overhead electrical fittings, and before taking up his present post he was manager of a research department which included a chemical laboratory, metallurgical and process control sections, and an operational research section. He later became manager of an operational research department. Mr Duckworth has published papers on powder metallurgy, fatigue, wear of metals, and on statistical and operational research subjects.

Professor G. Wesley Austin, O.B.E., M.SC., F.I.M., will continue to act as consultant to the division.

Industrial training in operational research

The introduction of a graduate apprenticeship scheme for specialized training in operational research was recently announced. The purpose of the scheme is to train recruits not only in the specialist techniques, but also in the practical art of applying these methods to actual industrial problems.

During the two-year course the apprentice will be regarded as a temporary member of BISRA staff at a salary of £700 per annum at the age of 21, with an extra £15 per annum for each year above that level. Training will consist of day-to-day experience as a member of a team engaged on actual problems in the steel industry followed by a more formal course in techniques, arranged in collaboration with the London School of Economics. The course will culminate in an examination for the Diploma in Operational Research recently instituted by the London School of Economics. The basic qualification for an apprentice is a good degree in a pure or applied science, mathematics, or statistics.

Further particulars, including an explanatory leaflet, can be obtained on application to BISRA's Personnel Officer at 11 Park Lane London W1.

Investigation proposed by Corrosion Advice Bureau

During the last twenty years the use of sprayed metal coatings of aluminium or zinc to protect structural steelwork has grown in importance, and this method of protection is now frequently specified by consulting engineers for structures to be erected both at home and abroad. So far as is known, satisfactory results have been obtained in the majority of cases, but recently there have been several instances of premature failure of the protective scheme in important structures treated in this way.

These observations have led to doubts regarding the merits of sprayed metal coatings beneath paint for the protection of structural steel. It is, therefore, important to clarify the position and lay down the conditions under which properly chosen and well applied composite protective schemes of sprayed metal coatings and paint can be relied on to protect steel satisfactorily for specified periods.

BISRA, through its Corrosion Advice Bureau, is taking the initiative in doing this, by organizing a collaborative investigation with the financial support of all who may be interested in the subject. Direct approaches have been made to the various trade associations and professional bodies concerned, but any others, including individual firms, who would like to participate are invited to write to the Secretary, Corrosion Advice Bureau, BISRA, 140 Battersea Park Road London SW11. It is envisaged that a total fund of £5000 will suffice for the purpose of the research, and the subscriptions already made or promised render it probable that this amount will be raised.

THE INSTITUTE OF METALS

Autumn meeting

The Autumn meeting of the Institute will be held in Bath from Monday to Friday, **5–9 September** 1960. The reception committee under the chairmanship of Mr D. M. G. Sneddon, general manager of the Imperial Smelting Corporation Ltd, Avonmouth, has arranged an attractive programme. There will be technical sessions on engineering properties of nickel-base alloys; lubricants for cold rolling aluminium; the relative merits of extrusion and other methods for the manufacture of bar, rod, and wire rod; and magnesium alloys for fuel-element components.

On the final morning the Autumn lecture on 'Metals and the world of engineering structures' will be delivered by Professor Sir Alfred Pugsley, O.B.E., D.SC., F.R.S.

Full details may be obtained from the Secretary, 17 Belgrave Square London SW1.

VEREIN DEUTSCHER EISENHUTTENLEUTE

Death of Dr.-Ing. A. Ristow

The death has been announced in Düsseldorf of Dr.-Ing. Arno Ristow, technical secretary of the various steelmaking committees of the VdEh.

Dr-Ing. Alfred Mund, of Fried. Krupp, Essen, has taken over business management of the steelmaking committees while remaining at Fried. Krupp.

CONTRIBUTORS TO THE JOURNAL

S. A. Lewitt, B.SC.(ENG.), M.I.E.E. – Chief electrical engineer, John Summers and Sons Ltd, Shotton (*Contributor to the July Journal*).

Sidney Alfred Lewitt joined the English Electric Company Ltd, Stafford, in 1931 as an engineering apprentice, and later transferred to the company's special engineering staff. He joined John Summers in 1937 as assistant electrical engineer, to work on the hot and cold strip mill development which was then starting. He was appointed to his present post in 1945.

S. A. Lewitt

Dr R. M. Goldhoff

R. M. Goldhoff, M.S., PH.D. – Applied research metallurgist, General Electric Company, New York. (*Contributor to the July Journal.*)

Robert M. Goldhoff was educated at the University of Cincinnati and the Ohio State University. After military service in the Second World War he joined the Dayton Malleable Iron Company as a foundry metallurgist. In 1950, after receiving his M.S. degree, he joined the Battelle Memorial Institute as research associate in physical metallurgy. He is at present carrying out research on high-temperature metallurgy with the Large Steam-Turbine-Generator Department of the General Electric Company. His recent published works concern extrapolation of short-time high-temperature data.

S. Yukawa, PH.D. – Applied research metallurgist, General Electric Company. (*Contributor to the July Journal.*)

Sumio Yukawa was awarded his PH.D. by the University of Michigan in 1954. He then joined the Materials and Processes Laboratory of the Large Steam-Turbine-Generator Department of GEC as metallurgist and was recently appointed supervisor, metallurgy applied research. He is principally concerned with problems of brittle fracture in alloy steels.

Dr S. Yukawa

Dr J. S. McMullin

J. G. McMullin, SC.D. – Supervisor of alloy steel research, Crucible Steel Company of America, Pittsburgh, Penn., USA. (*Contributor to the July Journal.*)

John G. McMullin was born in Huntingdon, West Virginia, and attended public schools there before entering the Massachusetts Institute of Technology where he received his B.S. degree in 1943, then entered the US Army. He returned to the MIT after the war and gained his further degrees. He was with the General Electric Company from 1952 to 1959, when he took up his present post with the central research laboratory of the Crucible Steel Company of America.

R. C. Rooney, A.R.I.C.—Research Staff, British Cast Iron Research Association, Alvechurch, Birmingham.

Ronald Charles Rooney was born in Birmingham in 1931 and educated at Saltley Grammar School and the College of Technology, Birmingham. He gained his A.R.I.C. and A.C.T.(BIRM.) in 1953. Since 1956 he has been on the research staff of the British Cast Iron Research Association working on the development of new analytical procedures, with particular reference to trace element analysis. Mr Rooney specializes in instrumental methods of analysis, and is a committee member of the Polarographic Society.

E. M. Summers

E. M. Summers—General Manager (Ironworks), The Stanton Ironworks Company Ltd, nr. Nottingham.

Edward Merrett Summers was educated at Kettering Grammar School, Wellingborough Technical Institute, and Leicester College of Technology. In 1930 he joined the Wellinboro' Iron Company Ltd, which was acquired by Stanton in 1932. After various appointments with the Stanton Mr Summers became blast-furnace manager at the company's Holwell works in 1940, and general superintendent of the works in 1949. He was appointed furnaces' general manager to the company in 1953 and took up his present position in 1958.

Mr Summers is a member of The Iron and Steel Institute, and a member of council of BISRA and chairman of the Ironmaking Panel.

EDUCATION

Courses at Borough Polytechnic

The Division of Metal Science, Borough Polytechnic, Borough Road London SE1, has announced some special evening courses for the winter term, 1960–61. These include: Recent developments in electrolytic metal finishing; Refractories: their manufacture, properties, and uses; Metallurgy of semi-conductors.

Another course, 'Corrosion and protection of metals' will be run on one afternoon and one evening each week throughout the session. At the end of it, students will be able to sit for the Polytechnic Certificate examination. Further details are available from Mr G. Isserlis at the Borough Polytechnic.

Post-doctoral fellowships in Australia

A small number of post-doctoral fellowships in the field of nuclear science and engineering will be awarded to British subjects holding the degree of PH.D. or equivalent. Fellowships will be tenable at any Australian university or with the Institute at Lucas Heights, NSW. Preference will be given to candidates whose research projects require use of the facilities available at the research establishment of the Australian Atomic Energy Commission. Initial tenure will be two years starting in the second part of 1960, and extensions to a total period of three years may be granted. Emolument will be not less than A£1 750 per year and funds for equipment and other expenses may be provided.

Nominations by universities or other institutions, or applications by qualified candidates will be received by the Institute up to 31 August 1960.

Further enquiries should be addressed to the Scientific Secretary, Australian Institute of Nuclear Science and Engineering, Private Mail Bag, Post Office, Sutherland, NSW, Australia.

National certificates and diplomas in metallurgy

A new scheme for the award of Higher National Diplomas in Metallurgy has been arranged by the Ministry in conjunction with The Iron and Steel Institute, the Institution of Mining and Metallurgy, The Institute of Metals, and The Institution of Metallurgists.

The diplomas will be awarded on the successful completion of approved three-year sandwich or full-time courses suitable for students who have had full-time education, including the study of chemistry, physics, and mathematics, up to the age of 18, or have been in regular employment for at least two years with concurrent part-time instruction up to Ordinary National Certificate standard in chemistry, physics, and mathematics.

Rules 111 (1955) which set out the arrangements and conditions for the award of National Certificates in Metallurgy have been revised to include particulars of the new scheme, and copies are available from the Ministry of Education, Curzon Street London W1.

FORTHCOMING CONFERENCES AND EXHIBITIONS

Technical and industrial exhibition at Charleroi

The fourth EITI (International Technical and Industrial Exhibition) will be held from 15 to 26 September 1960 in the Palais des Expositions, Charleroi, Belgium. This year it has been decided to devote the exhibition exclusively to specialized industrial and technical products.

In conjunction with the exhibition, the Association Belge pour Favoriser l'Étude des Verres et des Composés Siliceux is holding a special 'journée' on minerals, agglomerates, and refractory materials. Further details are available from the offices of the Association at 13 Rue des Poissonniers, Brussels.

Symposium on automatic control

The Institution of Mechanical Engineers is arranging for the British contribution to the First International Congress of the International Federation of Automatic Control, held in Moscow from 27 June to 7 July, to be discussed at a symposium in London on 27 and 28 September 1960. Besides discussing the 29 British papers it is planned to report briefly the more important contributions to the Moscow congress.

Preprints of the British papers are available and will be supplied to those registered to attend the symposium. A list of the papers together with registration forms is available from the Secretary, The Institution of Mechanical Engineers, 1 Birdcage Walk London SW1.

Conference on clean air

The 27th annual conference of the National Society for Clean Air will be held at Harrogate from 5 to 7 October 1960. The conference will include two sessions for the presentation and discussion of reports on the inquiry into the working of the Clean Air Act now being made by the Society by means of a questionnaire to all authorities. There will also be an address by Sir Keith Joseph, M.P., Parliamentary Secretary to the Minister of Housing and Local Government, the presidential address by Sir Hugh Beaver, and the Des Voeux Memorial Lecture by Sir Linton Andrews. The Clean Air Exhibition which will be held in conjunction with the conference will be of considerable practical value and interest to all delegates.

Registration forms and further details are available from the secretary of the Society at Palace Chambers, Bridge Street London SW1.

Effluent and water treatment exhibition and convention

What is believed to be the first exhibition and convention of its kind is to be held at the Seymour Hall, Seymour Street London W1, from 18 to 21 October 1960. Papers to be presented at the convention include subjects such as biological treatment of industrial wastes, river pollution prevention in Scotland, pumping, the application of polarography to the analysis of sewage and industrial wastes, problems of design in effluent treatment plants, coastal pollution, disposal of radioactive effluent, and water softening for industrial and domestic purposes. Further details are obtainable from P. I. Craddock, Press and Information Officer, Thunderbird Enterprises Ltd, 140 Cromwell Road London SW7.

Australian symposium on phase transformations

A symposium on 'Phase transformations in metals and alloys' will be held at the University of Melbourne Engineering School on 10 and 11 November 1960. This symposium is being sponsored by the physical metallurgy division of the Melbourne branch of the Australian Institute of Metals. Thirteen papers will be presented which will be published in the August 1960 issue of the *Journal of the Australian Institute of Metals*.

Corrosion and Metal Finishing Exhibition

Some of the country's leading corrosion scientists and engineers will take part in a meeting being held in connexion with the Corrosion and Metal Finishing Exhibition at Olympia this autumn. The Corrosion Group of the Society of Chemical Industry will be holding the meeting on Thursday, 1 December, in the Empire Restaurant, Olympia, at the invitation of the organizers of the exhibition. Papers to be presented will probably deal with: 'The influence of water movement on corrosion.'

The Corrosion and Metal Finishing Exhibition will open at 11 am on Tuesday, 29 November and will remain open until Friday, 2 December 1960.

Full details can be obtained from the Exhibition Director, 9 Eden Street London NW1.

NEWS OF SCIENCE AND INDUSTRY

Changes at Mond Nickel and Henry Wiggin

The following organizational changes in The Mond Nickel Company Ltd and its subsidiary company, Henry Wiggin and Co. Ltd, became effective from 1 July.

Mr G. ARCHER relinquished his position as chairman of Mond and Wiggin and was appointed president of Mond and Wiggin.

Mr IVON A. BAILEY (*Member*) was appointed chairman and chief officer of Mond and Wiggin.

Dr L. B. PFEIL (*Member of Council*) was appointed vice-chairman of Mond.

Mr J. O. HITCHCOCK (*Member*) was appointed managing director of Mond and deputy chairman of Wiggin.

Mr H. W. G. HIGNETT (*Member*) was appointed managing director of Wiggin.

Dr G. L. J. BAILEY relinquished his position as superintendent of Mond's development and research laboratory in Birmingham and transfered to London to become manager of research.

Mr E. J. BRADBURY (*Hon. Member of Council*) succeeded Dr Bailey as superintendent of Mond's development and research laboratory in Birmingham.

New plant and equipment

Foundry Equipment Ltd recently announced a centrifugal casting machine which allows the production of liner castings 30% lighter than by conventional methods. Increased production and higher quality castings are also claimed for the machine, which has created wide interest both in the UK and abroad.

Sharston Engineering Co. Ltd of Wythenshawe, Manchester, have been appointed sole UK distributors for the Italian 'Valdarno' trimming press. The main use of the press is to trim off 'flash' and other excess metal from hot or cold steel castings and forgings, and it is said to be particularly suitable where a high throughput is required. The press frame is designed to allow for modification to the table size or form, to suit special requirements of the user.

Low-cost computer

What is described as a 'low-cost approach to computer control of processes' has been made by de Havilland Propellors Ltd in the 'Anatrol' computer system for the process industries, which was recently announced. The makers claim that the computer can be integrated into existing control schemes easily and efficiently and that it has a computing accuracy compatible with the accuracies of present-day measurement techniques and a computing potential comparable with that of a more costly and complex computer system.

Continuous casting plant for Spanish steelworks

A vertical continuous casting plant was recently put into operation at the Industrias del Besos S.A., Barcelona, Spain. This plant was built by Schloemann Aktiengesellschaft, Düsseldorf, in conjunction with Concast Aktiengesellschaft, Zürich. It is designed for the production of carbon steel billets in sizes from 2 in. square to 4 in. square and 8 in. × 2¾ in. slabs and ladle capacity is 15 tons. After a few days of running-in, operation of the plant was taken over by a Spanish team.

The industry made plain

An interesting booklet '*Iron and steel–master of them all*' has been published by Head Wrightson and Co. Ltd. Written by the former chairman of the Head Wrightson group, Mr Richard Miles, the booklet was originally intended for the group's own apprentices, but its appeal is such that it has had a wider distribution. It is a brief history of iron and steel and a survey of the industry in the UK at present, simply written and sustaining interest throughout. Line and half-tone illustrations combine instruction and light relief.

Copies of the booklet are available from the publicity office of the company at 20 Buckingham Gate London SW1.

Contract news

Head Wrightson Stockton Forge Ltd, a subsidiary of Head Wrightson and Co. Ltd, is supplying to Richard Thomas and Baldwins Ltd six rotary dryers for an iron-ore drying plant costing about £240000.

Three of these machines, each of which measures 45 ft long by 10 ft dia. will be installed at the Redbourne works and three at

the Spencer works of RTB. Each installation is capable of handling a weekly output of 21350 tons of ore processing from 0 to 2½ in.

John Lysaght's Scunthorpe Works has placed a £2½ m. order with the *Wellman Smith Owen Engineering Corporation* in collaboration with *VOEST* of Austria for the design and supply of a complete LD–AC oxygen steelmaking installation at Scunthorpe. The plant will include a 1000-ton hot-metal mixer, two 50–60-ton converters, cranes and handling machines, waste-heat boilers, gas-cleaning plant, buildings, and ancillary equipment. It is designed initially for the production of 400000 tons of ingots per annum, with provision for extension to 800000 tons.

Morgan Refractories Ltd of Neston, Cheshire, a member of the Morgan Crucible group of companies, has sold a substantial quantity of MR2 super duty firebricks to line part of the stack of an Australian blast-furnace. The order is being shipped to The Broken Hill Proprietary Co. Ltd, and is worth about £25000.

Instructions to proceed with work on gasholder contracts with a value of about £1m. have recently been received by *P.G. Engineering Ltd* from clients in the UK and overseas. The most recent contract received is for a 3m. ft³ A.B.P.–Klonne type gasholder to be installed at John Lysaght's Scunthorpe Works Ltd. This holder will store blast-furnace gas and is planned for completion by the end of 1961.

Air Products (Great Britain) Ltd announce that it is building a 380 tons/day oxygen plant at the O. Sinigaglia steelworks of Cornigliano, Genoa, Italy. The order for the plant, valued in excess of £1m., was secured despite strong competition from Germany, France, and the UK. The plant, worth more than £1m., the first to be built for an Italian steel company, will consist of two generating units each capable of producing 190 tons/day of oxygen and storage for 560 tons of liquid oxygen, and 420000 ft³ of high-pressure gaseous oxygen will also be included.

Change of address

The London office of Hadfields Ltd is now at 25 Berkeley Square W1 (HYD 0431).

INDUSTRIAL PUBLICATIONS RECEIVED

Wiggin Nickel Alloys no.56 includes articles on eddy current testing, the use of Nimonic 80A in valves for the Vanwall racing car, and two seagoing applications of Monel alloy in trawlers and runabout pleasure craft.

From Hubert Morris Ltd comes book no.242, describing a comprehensive range of the company's conveyors and some interesting handling schemes.

Plenty and Son Ltd have issued a leaflet describing the complete range of Plenty Duplex filters.

A leaflet from Materials Handling Equipment Ltd describes the 'Traveloader' heavy-duty side-operating fork-lift transporters.

Catalogues BA.73 and BPV.3 from Turner Brothers Asbestos Co. Ltd describe the company's rubber conveyor belting and 'Poly-V' drives respectively.

Northide Ltd give a brief description of their new safety goggle made entirely of nylon and weighing only 1 oz.

'Radiant jet furnaces' is the title of leaflet V61 from the Incandescent Heat Co. Ltd.

The April issue of *Welding News* from Quasi-Arc Ltd includes articles on welding the 'Oriana' and flux materials for electrode coverings.

A finely produced booklet 'Nitrogen and life' comes from La Société l'Air Liquide, France.

The *Bulletin* for May, issued by Baird and Tatlock (London) Ltd details the company's products and their applications.

A booklet 'The ABC of electronic brains' has been issued by Elliott–Automation Ltd.

A new protective coating, Tretol Coal Tar/Epoxy 177 is described in a recent leaflet from Tretol Ltd.

Leaflet V.4.B from Thompson L'Hospied and Co. Ltd lists standard sizes of the company's cast nickel–chromium circular carburizing pots and rectangular carburizing boxes.

Two revised brochures (nos.12 and 14) from Samuel Osborn and Co. Ltd cover the company's range of milling cutters and reamers.

'Data sheets on dust collectors' is a new book from the British Steel Castings Research Association.

DIARY

30 July 7 Aug. SWEDISH INDUSTRIES FAIR—Malmo, Sweden.

5–14 Aug. INTERNATIONAL TRADE AND EXPORT FAIR—Dornbirn, Austria.

15–24 Aug. INTERNATIONAL UNION OF CRYSTALLOGRAPHY—Conference and symposia—Cambridge.

15–25 Aug. INTERNATIONAL GEOLOGICAL CONGRESS—Stockholm.

16 Aug. THE INSTITUTE OF METALS—Lecture, *The physical nature of plastic flow*, by Dr J. J. Gilman (Research Laboratory, GEC, Schenectady, USA)—17 Belgrave Square London SW1.

31 Aug.–7 Sept. BRITISH ASSOCIATION FOR ADVANCEMENT OF SCIENCE—Conference—Cardiff.

4–11 Sept. INTERNATIONAL AUTUMN TRADE FAIR—Vienna.

5–9 Sept. INSTITUTE OF METALS—Autumn meeting—Bath.

6–16 Sept. PRODUCTION ENGINEERING SHOW—Chicago, USA.

9–10 Sept. CZECHOSLOVAK SCIENTIFIC-TECHNICAL ASSOCIATION—Symposium on fatigue strength—Prague.

10–17 Sept. INTERNATIONAL INSTRUMENTS EXHIBITION—Stockholm.

10–25 Sept. GERMAN INDUSTRIES EXHIBITION—Berlin.

11–20 Sept. GERMAN MACHINE TOOL FAIR—Hanover.

11–25 Sept. INTERNATIONAL TRADE FAIR—Brno, Czeckoslovakia.

15–18 Sept. BRITISH WIRE ROD ROLLERS ASSOCIATION—Conference—Scarborough.

15–26 Sept. INTERNATIONAL INDUSTRIAL AND TECHNICAL EXHIBITION—Charleroi, Belgium.

17 Sept.–2 Oct. INTERNATIONAL TRADE FAIR—Mons, Belgium.

21 Sept.–1 Oct. FACTORY EQUIPMENT EXHIBITION—Belle Vue, Manchester.

22–24 Sept. INSTITUTE OF PHYSICS *et al.*—Conference on *Theory and practice of ultrasonic inspection*—Queens Hotel, Cheltenham.

26–28 Sept. FEDERATION OF BRITISH INDUSTRIES—Conference—Palace Hotel, Buxton.

26–30 Sept. INTERNATIONAL INSTRUMENT-AUTOMATION EXHIBITION—Pittsburgh, Penn., USA.

27–28 Sept. THE INSTITUTION OF MECHANICAL ENGINEERS—Symposium on automatic control—1 Birdcage Walk London SW1.

27–29 Sept. DSIR—Conference on ergonomics—Connaught Rooms, London WC2.

BRITISH IRON AND STEEL INDUSTRY TRANSLATION SERVICE

The following translations are now available, in addition to those given on page 344 of the July, 1960 issue of the *Journal:*

Fundamental principles on the digital presentation and processing of data with special reference to computers.
WARTMANN, R., *Arch. Eisenhüt.*, 1959, Aug., pp.513–518. (£3 15*s.* 0*d.*) **[1464**

Discussion of paper on 'Research into the formation and elimination of deoxidation products', by E. Plöckinger and M. Wahlster (BISI No.1544).
BORN, K., *Stahl u. Eisen*, 1960, May 12, p.669. (£1) **[1544A**

The sulphur equilibrium between carbon-saturated hot metal and lime-silica slags.
SCHENCK, H., *et al.*, *Arch. Eisenhüt.*, 1960, Jan., pp.11–17. (£4 5*s.* 0*d.*) **[1641**

Improvement in the production of open-hearth furnaces heated by low calorie gas by the injection of compressed fluids into their gas burners.
SEPTIER, L., *et al.*, *Rev. Mét.*, 1959, Feb., pp.179–202, and *Centre Doc. Sidér. Circ.*, 1959, (2), pp.445–460. (£6) **[1693**

The distribution of the burden within the blast furnace and some methods of affecting it.
HEYNERT, G., *et al.*, *Stahl u. Eisen*, 1960, Apr.14, pp.473–483. (£7 5*s.* 0*d.*) **[1716**

Blast furnace slags as a commercial product.
KIELSKI, P., *Hutnik* (Poland), 1951, (6), pp.225–227. (£2 15*s.* 0*d.*) **[1733**

Road brick made out of Mansfeld slag.
WAGENMANN, K., *Hutnik* (Poland), 1951, (6), pp.229–233, and *Technik*, 1950, (4), pp.170–172. (£4 10*s.* 0*d.*) **[1734**

Influence of organic compounds in ceramic mixes. A technological study of the fabrication of porous materials.
MANN, W., *Deutsche Keram. Ges. Berichte*, 1960, Jan., pp.11–22. (£5) **[1735**

Semi-automatic manipulator for closed-die forging hammers.
ALBERT, M. A., *Kuznechno-shtampovochnoe proizvodstvo*, 1959, (7), pp. 38–40. (£2 15*s.* 0*d.*) **[1738**

Cold-rolled dynamo strip.
SCHLUTER, H., and F. STABLEIN, *ETZ-A*, 1959, Sept.1, pp.576–582. (£4 10*s.* 0*d.*) **[1740**

Use of ultrasonics for seam and spot welding.
OL'SHANSKII, N. A., *et al.*, *Avtom. Svarka*, 1958, (10), pp.75–80. (£3 5*s.* 0*d.*) **[1747**

Creep and rupture of tubes under internal pressure.
KATS, S. N., *Izv. Akad. Nauk S.S.S.R., Otd. Tekh. Nauk*, 1957, Oct., pp.86–89. (£2 5*s.* 0*d.*) **[1748**

A new optical method for the determination of non-metallic inclusions in steel.
ERNST, T., and G. KÖNIG., *V.D. Emailfachleute Mitt.*, 1959, Dec., pp.93–96. (£3) **[1752**

The level of development of the all-basic open-hearth furnace in the German Democratic Republic (slightly abridged).
RABES, F., and L. ULLSPERGER, *Neue Hütte*, 1960, Apr., pp.224–232. (£5 15*s.* 0*d.*) **[1759**

New procedure for heating open-hearth furnaces with cold gas and fuel oil.—I.
SCHMITT, W., *Stahl u. Eisen*, 1960, Apr.28, pp.616–619. (£3 5*s.* 0*d.*) **[1760**

Research into the influence of addition elements in austenitic steels on the limit of elasticity at room temperature.
NOHL, H., *et al.*, *Mém. Scientifiques Rev. Mét.*, 1959, (1), pp.22–29. (£4) **[1783**

On the location of modern iron works at the coast.
INGEN HOUSZ, A. H., *Stahl u. Eisen*, 1960, March 3, pp.272–276. (£2 15*s.* 0*d.*) **[1784**

1791–1793 *Papers presented at a colloquium on deep-drawing held in Paris, 23–25 May* 1960.

The influence of various factors on the results of the Erichson Test.
JENTET, M. (£3) **[1791**

Ultrasonic testing of deep drawing sheets.
LEHFELDT, W. (£3 10*s.* 0*d.*) **[1792**

Effect of the anisotropy of sheets on their plastic deformation under uniaxial and biaxial stresses.
CRUSSARD, C., *et al.* (£5 10*s.* 0*d.*) **[1793**

1803 *and* 1804 *Papers presented at the international discussions on long time behaviour of high temperature steels held in Düsseldorf, 24–25 June* 1960.

Metallographic investigations on X 8 CrNiNb 1613 after creep tests at 750°C.
SCHRADER, A., and A. KRISCH. (£6 10*s.* 0*d.*) **[1803**

Scatter of creep strength of steel X 22 CrMo (W) V 12 1, a joint evaluation of German tests.
SCHINN, R. (£3 5*s.* 0*d.*) **[1804**

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects

CONTENTS

Note: The letters *a*, *b*, and *c* refer to the three columns on the page concerned.

	PAGE
MINERAL RESOURCES	479*a*
ORES—MINING AND TREATMENT	479*a*
FUEL—PREPARATION, PROPERTIES AND USES	480*b*
AIR POLLUTION AND SMOKE	480*b*
TEMPERATURE MEASUREMENT AND CONTROL	480*c*
REFRACTORY MATERIALS	480*c*
IRON AND STEEL, GENERAL	481*a*
BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON	481*b*
PROPERTIES, TREATMENT AND USE OF SLAGS	482*b*
DIRECT PROCESSES	482*b*
PRODUCTION OF STEEL	482*c*
FOUNDRY PRACTICE	484*a*
VACUUM METALLURGY	487*b*
REHEATING FURNACES AND SOAKING PITS	488*a*
HEAT-TREATMENT AND HEAT-TREATMENT FURNACES	488*b*
FORGING, STAMPING, DRAWING, AND PRESSING	489*b*
ROLLING MILL PRACTICE	491*a*
MACHINERY AND SERVICES FOR IRON AND STEELPLANT	492*b*
WELDING AND FLAME-CUTTING	493*a*
MACHINING AND MACHINABILITY	494*a*
CLEANING AND PICKLING	494*c*
PROTECTIVE COATINGS	495*c*
CLAD SHEET AND HARD FACING	497*a*
POWDER METALLURGY	497*b*
FERRITES, CERMETS AND CARBIDES	498*a*
PROPERTIES AND TESTS	498*a*
METALLOGRAPHY	505*c*
CORROSION	507*a*
ANALYSIS	510*a*
INDUSTRIAL USES AND APPLICATIONS	511*a*
HISTORICAL	511*c*
ECONOMICS AND STATISTICS	511*c*
MISCELLANEOUS	512*b*
BOOK NOTICES	512*a*
NEW PUBLICATIONS	515*a*

* Abstracts, the reference to which is marked with an asterisk, are or are to be made available as translations.

ABSTRACTS

These abstracts are also available on index cards in advance of publication in the *Journal*, each abstract being classified under the Universal Decimal system (UDC). Details and subscription notes can be obtained from the Secretary, The Iron and Steel Institute.

MINERAL RESOURCES

Salzgitter brown iron ores: basis for a second Ruhr J. B. Huttl (*Eng. Min. J.*, 1959, **160,** Nov., 82–86) The location, extent, properties, and treatment of the Salzgitter deposits are described.

The rationalization of the Siegerland ore mining industry G. Schröder (*Stahl Eisen*, 1960, **80,** May 12, 652–659) The re-organization and mechanization of the Siegerland iron ore mining industry is briefly described. This process was started in 1953 and the first results are now becoming evident. A battery of sintering furnaces has been erected.—T.G.

Remarks on a paper by E. Schubert, H. Bayer and E. Schwarz on study of the red-iron ore deposits in the Eifel K. C. Taupitz (*Erzmetall*, 1960, **13,** April, 170–174) Taupitz supplements the account by Schubert *et al.* from his own studies and presents some charts. Schubert replies.—T.G.

Materials for Russian steel (*Min. J.*, 1960, **254,** April 15, 431) A note, mainly on ore reserves.

The iron ore deposit of Marquesado (Spain) and methods of mining it M. G. Rozière and L. Lamana de Hoyos (*Rev. Ind. Min.*, 1959, **41,** Nov., 897–932) Methods of open-cast and underground mining, screening and loading, and output and productivity of this Sierra Nevada mine are described. The article concludes with a description of the iron ore resources of Spain.

Arsenical iron ores as raw material for the ferrous metallurgy P. Dumitrascu and M. Ispas (*Met. Constr. Masini*, 1959, (11), 928–933) The Kerch iron ore deposits in the USSR contain about 0·1%As but in view of the large reserves available and their proximity to Rumania, experiments are being carried out for concentrating these poor ores and extracting the As. A few alternative methods are being tried and the fluidization of the ore layer shows considerable promise.

ORES—MINING AND TREATMENT

A review of the ore industry G. J. Holt (*Blast Furn. Steel.*, 1960, **48,** Jan., 40–44) American ore production is reviewed with particular emphasis on recent developments. The American iron industry depended mainly on Lake Superior ore but this dependence is now changing and importation of iron from Canada and abroad is making heavy inroads into the former markets for Lake Superior ore.—T.G.

Blasting with deep holes in open mines B. Kolaczkowski and J. Godyn (*Rudy i Metale*, 1959, **4,** (4), 115–119) As blasting with shallow holes in open mines is uneconomical, deep holes have been used since 1957. The authors describe the regulations governing such work and also methods and equipment used. The prospects of success in such deposits as the 'Boleslaw' ores, using deep holes, are very good.

Narvik—the modern ore harbour (*Min. J.*, 1960, **254,** April 22, 462–463) A brief account of equipment.

Narvik—Europe's largest ore-handling installation K. Schmoll (*Chem. Eng. Min. Rev.*, 1959, **52,** Oct. 15, 63–65).

St. Lawrence seaway offers new avenue for output of Canadian and U.S. mines (*Eng. Min. J.*, 1959, **160,** Aug., 82–83, 86–88) With map and a list of ore districts and producers of iron ores and other minerals.

Formation of metal sulphides and oxides during the thermo-chemical decomposition of manganese and iron sulphates V. V. Pechkovskii (*Zhur. Prikl. Khim.*, 1959, **32,** (12), 2613–2618) The experimental findings here described have established that the velocity and the degree of formation of sulphides during the reduction of the sulphates in the presence of carbon depends on temp., composition of the reducing agents, degree of sulphidization of the oxides which are formed from the sulphur-containing products of the reduction processes, and on the concentration of oxygen and SO_2 in the gaseous phase.

New iron ore preparation plants put into operation at the Ferrière-aux-Etangs mine (western coalfield) R. Lucas and P. Sauvage (*Rev. Ind. Min.*, 1959, **41,** Nov., 975–984) A detailed description of the screening and storage plants of this roasted blackband ore.

Fundamental study on the removal of arsenic from limonite during the ore-beneficiation process. II. Mechanism of the removal of arsenic from limonite and accelerating materials for arsenic removal S. Sato (*Tetsu to Hagane*, 1959, **45,** Oct., 1131–1135) The relevant equations are $4\,FeAsO_4 = 2\,Fe_2O_3 + 2O_2 + As_4O_6$ and $4\,FeAsO_4 + 4\,CO = 2\,Fe_2O_3 + 4\,CO_2 + As_4O_6$, of which the latter is more rapid. Coal, coke, and pyrites accelerate As removal, but $FeCO_3$, $MnCO_3$, and $FeCl_2$ are not effective.—K.E.J.

Fundamental study on the removal of arsenic from limonite during the ore beneficiation process. III. Effect of gangue components of mixed iron ores on the removal of arsenic from limonite S. Sato (*Tetsu to Hagane*, 1959, **45,** Nov., 1231–1236) Removal of As from Hokkaido and Dungun ores during sintering after ore blending is hindered by basic gangue constituents (CaO, MgO) but hardly affected by neutral (Al_2O_3) or acidic (SiO_2, TiO_2) materials. The major effect was that of CaO at 1300°C.

Physico-chemical basis of the dearsenication of the Kerch iron ores P. T. Danil'chenko, V. F. Kovtun, and A. G. Lagutina (*Izv. AN Metallurg. i. Topl.*, 1959, (4), 8–14) Both arsenates and iron arsenites, $FeAsO_4$. $4H_2O$, $Fe_3(AsO_4)_2$. $8H_2O$, and $FeHAsO_3.nH_2O$, when heated in air or in a mixture with carbon in a current of water vapour or CO_2 just like Fe_2As when heated in O_2, CO_2 of H_2O give up the As in the form of As_2O_3, which leaves the reaction zone. The reaction products are iron oxides mixed with 0·5%As. Laboratory experiments have shown that the Kerch ores, with the original structure can be dearsenicated almost completely when heated at between 1000 and 1050° in a current of water vapour, retaining only traces of As (0·005–0·01%). Optimum operating conditions have been worked out for this process on an industrial scale.

The formation of the magnetic fraction on roasting Fe–Mn compounds B. Hadáček and J. Hruby (*Hutnik. Listy*, 1959, **14,** (11), 963–970) The formation and percentage content of magnetic fractions in mixtures of Fe_2O_3 and various Mn oxides and carbonates roasted at temp. of 600–1100°C in N_2, air, SO_2, or with carbon additions, were investigated. Best results were obtained with $MnCO_3$; optimum yields, i.e. conversion of Mn into the magnetic phase, was obtained with Mn:Fe ratios of 1:2 to 1:1.

How Salzgitter brown iron ore is upgraded to blast furnace feed J. B. Huttl (*Eng. Min. J.*, 1959, **160,** Nov., 86–92) Beneficiation of the lean high-Si ore at the Calbecht plant of Salzgitter AG is described.

High-intensity magnetic separation of iron ores O. E. Palasvirta (*Min. Eng.*, 1959, **11,** Dec., 1244–1248) Theory, design, and operation are described.

Methods of producing high quality iron ore concentrates from poor magnetite ores S. G. Evsiovich (*Izvest. AN Metallurg i Topl.*, 1959, (3), 7–12) From poor Krivoi Rog magnetite ores, it is possible to obtain concentrates containing 65–66% iron and 7–8% silicon by multi-stage methods of enrichment with five to seven successive operations in which the granulation of the ore decreases to 0·074 mm from 25 to 40% in the first stage to 90–95% in the last. These methods are most economical, with classification and enrichment in a closed cycle which includes also pulverization, and are now used industrially with the reconstruction of old and the erection of new enrichment plants.

Manganese responds to dithionate (*Eng. Min. J.*, 1959, **160,** Nov., 96–98) A description of the

process by which low-grade Mn ore in the Arizona deposits is upgraded by suspending the pulverized ore in a solution of calcium dithionate and leaching.

Investigation into recrystallization of powdered iron ores A. Ishimitsu, T. Furui, and K. Sugawara (*Tetsu to Hagane*, 1959, **45**, Sept., 915–917) The effects of heating FeO, Fe_2O_3, Fe_3O_4, and mixtures of these with and without slag in various atmospheres are examined.

On the selective chlorination roasting of pyrites cinder in the pilot plant fluidizing roaster K. Okajima, M. Inouye, K. Sano, and Y. Sugiura (*Tetsu to Hagane*, 1959, **45**, Sept., 897–899) Data are given for the reduction of Cu, S, and Zn contents of four types of cinder.

Fundamental study of the sintering of iron ores. Influence of air flow, coke content, and moisture content on the temperature distribution of the sintering bed K. Sanbongi, N. Nishida, and A. Suzawa (*Tetsu to Hagane*, 1959, **45**, Sept., 924–926) Bed temp. distributions are given for charges of Larap ore, with varying sintering parameters.—K.E.J.

Effect of coke size on the productivity of iron ore sintering K. Miyagawa (*Tetsu to Hagane*, 1959, **45**, Sept., 922–924) Sintering times are generally lowest with breeze between 9–16 and 16–32 mesh, productivity highest with slightly higher sizes, and strengths highest with lower sizes.—K.E.J.

Production of fluxed sinter from Kerch tobacco-ore concentrate Z. I. Nekrasov, N. A. Gladkov, and A. G. Ul'yanov (*Stal'*, 1960, (4), 293–298) Sintering trials are reported and optimum conditions, particularly moisture content, were established.

What's needed for improving your sintering setup R. Houlton (*Iron Age*, 1959, **184**, Sept. 17, 106–108) A review is made of automation controls available for sinter plants. These include X-ray emission analysis of raw materials, feed regulators, and the linkage of flow of mix to the strand with the position of the burn-through point.—A.G.

Control of the basicity of self-fluxing sinter U. Fujii, T. Numata, R. Ando, and T. Murakami (*Tetsu to Hagane*, 1959, **45**, Sept., 911–913) Specific ores and blends are evaluated in terms of their lime-equivalence.—K.E.J.

Study on the improvement of the quality of self-fluxing sinter T. Doinouchi, A. Chida, and S. Obuchi (*Tetsu to Hagane*, 1959, **45**, Sept., 920–922) The effects of varying contents of lime, coke, OH slag, and roll scale are studied.

On the matrix in self-fluxing sinter S. Watanabe and T. Matsuoka (*Tetsu to Hagane*, 1959, **45**, Sept., 913–915) Mineral constituents are identified by their refractive indices, and examined in terms of the CaO–MgO–Al_2O_3–SiO_2 quaternary diagram.—K.E.J.

Studies of characteristics of a cooling bin for sintered ore R. Aoyagi, Y. Saito, S. Ota, and M. Watanabe (*Tetsu to Hagane*, 1959, **45**, Sept., 917–919) Operating results from the Kamaishi works are given, and the effects of air vol. and travelling time on cooling characteristics are shown graphically.—K.E.J.

Fully automatic bulk handling lubrication system for a sinter plant G. Williams (*Iron Coal Trades Rev.*, 1960, **180**, April 1, 747–751) An illustrated account of the installation at the Margam Works.

Work on the production of high-basicity sinter when sintering fine concentrates N. M. Yakubtainer, E. V. Nevmerzhitskii, and G. F. Grigor'evykh (*Stal'*, 1960, (3), 193–204) The composition of Olenegorsk concentrate over a period is shown and developments at Cherepovets enabling the furnaces to run without added limestone and with the best coke rate in the USSR are described. Improvements to reduce the cost of sinter are foreshadowed.

The sintering of dust in fluidised-bed roasting N. Lowicki (*Z. Erz. Met.*, 1959, **12**, Oct., 497–507) Investigations on the Fe–O–S system in the region of the 1000° and 1300°C isotherms showed that dust sintering in the furnace takes place mainly in particles up to 0·2 mm dia. An essential factor in the formation of Fe oxide–FeS compounds melting at 1000–1100°C is a S content between 18 and 30% (22 refs).

Tests on a fluidised-bed furnace for the magnetic roasting of iron ores V. I. Karmazin, K. E. Makhorin, G. V. Gubin, A. M. Kucher, and V. E. Gasparyan (*Stal'*, 1960, (4), 289–293) Trials of concentration processes on low-grade oxidized ores are described. The apparatus is described and gas consumption, output, and degree of concentration are reported. The yield was 55% of a fraction containing 65–66%Fe and a recovery of 86–88% of the Fe in the ore.

Fundamental study on the reduction of iron ore in fluidized beds T. Akutagawa and A. Okura (*Tetsu to Hagane*, 1959, **45**, Sept., 895–897) Curves show the effect of temp. (500–700°C) on the reduction of Goa hematite by H_2.—K.E.J.

The production of iron coke C. Eymann (*Stahl Eisen*, 1960, **80**, May 12, 684–686) A review of the work carried out in Poland and Czechoslovakia on the production of iron coke and its use, for instance, its effect on the lining of the blast-furnace.—T.G.

The production and use of ferro coke G. Granger, M. J. Lach, and J. Mitchell (*Coke Gas*, 1960, **22**, April, 147–154) From the AIMME Blast Furnace Conference at St Louis.

Reduction of iron ores by a mixture of petroleum gas and water vapour M. Z. Enikeeva, S. S. Krivolapov, P. D. Obolentsev, and V. P. Rozhdestvenskii (*Izves. AN Metallurg i Topl.*, 1959, (4), 32–41) The best operating conditions for the reduction of the Tukansk and Orsk iron ores by mixtures with petroleum gas and water vapour are: temp. 900–950°; ratio vapour/gas = 1; rate of gas supply 6 l/h, and of petroleum gas 4·5–5·1 for one passage of 30 g Tukansk ore, or 3·5–4·1 for 30 g of Orsk ore. Under these conditions the transformation of the iron oxides into metallic iron proceeds to 95–96% for the Tukansk ores and to 90–91% for Orsk while the carbon content in the reduced ores does not exceed 1·0–1·3 and 1·5–2·0% respectively.

FUEL—PREPARATION, PROPERTIES AND USES

A graphical method of determining optimum operating conditions for washing mixed coals G. G. Sarkar, S. Sakha, and A. Lahiri (*J. Inst. Fuel*, 1960, **33**, March, 145–150).

New aspects of the question of the influence of crushing and mixing plants for coking coal on the quality of the coke A. Jenker (*Glückauf*, 1952, **88**, 363–367; *BCRA translation* 26, 1952, Oct., pp.7).

Moisture and size effect on bulk density of compressed mixtures of crushed coal charged in coke ovens G. Odone and F. Ambrosetti (*Coke Gas*, 1959, **21**, Nov., 441–445, 480).

Correlation and utilisation of results from the laboratory and coal processing. B. High-temperature coke, in the laboratory and in practice W. Mantel (*Brennstoff-Chemie*, 1950, **31**, 161–172; *BCRA translation* No.14, 1952, Aug. 14, pp.8).

Study on manufacturing conditions of the coke used for a 1500-ton blast furnace. II. Proper conditions of coke-making with respect to the coal supplied to blast furnaces of the Yawata Iron Works H. Joh and S. Ida (*Tetsu to Hagane*, 1959, **45**, Sept., 887–889) Properties of coke made from given ratios of strongly and weakly coking coals are detailed.—K.E.J.

A steam plant for oil dehydration O. N. Bagrov and B. N. Khramov (*Stal'*, 1960, (3), 283–284) A description of the fuel oil drying plant at Severskand its advantages over OH waste gas heated installations.

The natural gas of the Lacq and the Northern Sahara deposits and its importance to the supply of fuel for Western Europe M. Ussar (*Stahl Eisen*, 1960, **80**, May 12, 676–684) The power demands of the individual Western countries are discussed with particular emphasis on the impact the new natural gas supply may have in the future.—T.G.

AIR POLLUTION AND SMOKE

The fight against air pollution in the iron and steel industry K. Guthmann (*Stahl Eisen*, 1960, **80**, April 28, 627–632) A comprehensive review of the efforts of the British iron and steel industry in its fight against air pollution.

Correction for anisokinetic sampling of gas-borne dust particles S. Badzioch (*J. Inst. Fuel*, 1960, **33**, March, 106–110).

Utilization of the smoke resulting from oxygen blowing M. Krijgsman (*Stahl Eisen*, 1960, **80**, April 28, 621–622) A plant is described for the precipitation of the brown fume that is formed in O_2-blowing. As the smoke contains up to 54%Fe, the sludge from the wet precipitation plant was made up with lime and added to fines for sintering in quantities of up to 10%. The economics of the utilization of sludge depend to a large extent on local conditions, mainly on the price of the fine ore.

Removal of dust in oxygen steelmaking. I S. Watanabe, S. Dohi, and Y. Ozawa. **II. The Venturi scrubber as a device to remove open hearth furnace dust** H. Kumai, Y. Ozawa, H. Hirayama, H. Kamei, and E. Takemura (*Tetsu to Hagane*, 1959, **45**, Sept., 935–937; 937–939).—K.E.J.

TEMPERATURE MEASUREMENT AND CONTROL

Recording optical pyrometer N. A. Blum (*Rev. Sci. Instr.*, 1959, **30**, April, 251–253) The range 1300–3000°C is covered with a time constant of a few millisec. It operates unattended and is specially designed for arc and plasma measurements.

Symposium on the optical pyrometer temperature scale (*Indust. Heat*, 1959, **26**, Oct., 2006–2012) A symposium at the NBS dealt with present temp. scales and practices, recent developments in optical pyrometry, and the possible use of the temp. scale beyond 4000°K.

The measurement of iron temperature in tapping and pouring E. Starke (*Giessereitechn.*, 1959, **5**, Aug., 237–240) The use of the colour-brightness radiation pyrometer the Bioptix and the Tricolor) and the immersion thermocouple are described and their advantages and drawbacks compared.

REFRACTORY MATERIALS

The supply of refractory products to the French iron and steel industry (*Aciers Fins Spéc.*, 1960, March, 4–10) The selection, supply, and treatment of various natural and processed refractories are discussed with regard to their application in various furnaces, mixers, converters, and ladles.—R.P.

Future trends in refractories C. S. Hedley (*Iron Steel*, 1960, **33**, May 25, (Special Issue), 272–275) A discussion of refractories for LD converters, OH furnaces, electric furnaces, soaking pits, and special purposes such as crucibles and hot tops.

Looking ahead for refractories W. F. Ford (*Claycraft*, 1959, **33**, Nov., 40–42, 44) A lecture.

Abrasion plates for use as mould walls in dry pressing of refractories P. Draignaut (*Sil. Indust.*, 1960, **25**, Feb., 65–70) The wear resistance of various materials in use as mould walls in the high pressure 'dry' extrusion of refractory products was studied. The best materials were found to be 12–13%Cr steels, or Cr–V steels. Hardened and tempered steels, as well as hard Cr steels were not found satisfactory. WC plates were most satisfactory and, in spite of their high cost, their use is recommended.—P.F.

The palletisation of refractory bricks (*United Steel Co. Ltd Report*, 1957, Oct., pp.17) Transport on British Railways is described.

Fundamental investigation of magnesite-chromite brick after service in the roof of a basic open-hearth furnace J. Oba and T. Hayashi (*Tetsu to Hagane*, 1959, **45**, Sept., 944–945) The variation of lattice parameters (of secondary ferruginous spinel and periclase) with distance from the hot face are plotted for used bricks.—K.E.J.

Heat-resistant concretes investigated (*Ceramics*, 1960, **11**, Jan., 30–31) N.B.S. work on four compositions is summarized.

Modernizing a firebrick works H. Pattinson (*Claycraft*, 1959, **33**, Dec., 66–70) Estimates of the economics of mechanization are made (in a Prize Competition essay).

Phase equilibrium studies in the system iron oxide–Al_2O_3–Cr_2O_3 A. Muan and S. Somiya (*J. Amer. Ceram. Soc.*, 1959, **42**, Dec., 603–613) The system has applications to chrome refractories in steel furnaces.

Thermal conductivity of magnesia from 1030–1880°C J. D. McClelland and E. H. Zehms (*J. Amer. Ceram. Soc.*, 1960, **43**, Jan.,

54) The thermal conductivity k of polycrystalline opaque MgO has been measured between 1030 and 1880°C. It was found that k can be expressed by $k = 18{\cdot}8/(T - 125)$. The radiation term used formerly by McQuarrie was not required.—T.G.

Apparent temperatures measured at melting points of some metal oxides in a solar furnace J. J. Diamond and S. J. Schneider (*J. Amer. Ceram. Soc.*, 1960, **43,** Jan., 1–3) Six metal oxides were melted in a solar furnace and their melting points determined by means of a conventional optical pyrometer. Al_2O_3, TiO_2, and Nb_2O_5 showed melting points in close agreement with those determined by other methods. Ta_2O_5, La_2O_3, and Dy_2O_3 varied widely and bore no relation to their melting points. The method is therefore not suited for the determination of unknown melting points.—T.G.

The phase rule and ceramics. IV C. Radford (*Ceramics*, 1959, **10,** Oct., 30, 32, 34, 36, 38) Three-component systems are outlined.

The Ateliers and Forges de la Loire. The Marais Works at St. Etienne (*Aciers Fins Spec.*, 1960, March, 82–92) A brief survey of the French iron and steel industry in the St Etienne area is given. This is followed by a description of the layout of OH and electric furnaces, roughing, rolling and finishing mills, soaking pits, and reheating furnaces as well as the control department. The plant produces semi-finished products, castings from 2 to 75 t, and various kinds of steel plate.—R.P.

IRON AND STEEL, GENERAL

Developments in the iron and steel industry during 1959 I. E. Madsen (*Iron Steel Eng.*, 1960, **37,** Jan., 93–145) Implications of the steel strike in the USA and its terms of settlement are discussed, and American developments in production of raw materials, iron and steel production and finishing, and in equipment during 1959 are reviewed in detail.

Visit to new plants of the HADIR works at Differdange (*Rev. Tech. Luxembourg*, 1959, **51,** (4), 233–236) Plants described briefly are: ore-crushing, new converter steel plant, strip mill, and tube works.

Activities of Verein Deutscher Giessereifachleute annual report 1958/9 P. Schneider (*Giesserei*, 1959, **46,** Oct. 8, 581–606).

Czechoslovakia's steel and metal industries (*Met. Bull.*, 1960, Feb. 12, i–iv, 21) A review of ore deposits, mining and smelting locations, and various other topics.

Metallurgical research behind the iron curtain (*Metallurgist*, 1960, **1,** March, 79–80) A brief account of impressions from visits to Czechoslovakia and Russia.

The report of the American Steel and Iron Ore Delegation's visit to the Soviet Union, May and June 1958 G. Heynert (*Stahl Eisen*, 1960, **80,** May 12, 688–696) This is an illustrated review of the official report by the AISI. The reviewer points out that he has refrained from making any comments on the findings reported.—T.G.

Eastern block states. Iron and steel industry (*European Coal and Steel High Authority Bibliography* 18, (1947–1959), 1959, May, pp.69).

Russia's seven-year plan outlines future steelmaking programme I. Denisenko (*Iron Steel Eng.*, 1960, **37,** Jan., 71–76) Iron and steelmaking targets are outlined, development of the eastern areas of the Soviet union is discussed, and plants for technical re-equipment reviewed.

Dominion Steel and Coal Corporation, Limited, initiates expansion programme (*Blast Furn. Steel Plant*, 1959, **47,** Aug., 818–836) The raw materials position, and the layout and operation of the blast furnace, steelmaking, rolling, and finishing sections and ancillary departments of the corporation are described and illustrated.

Problems associated with the setting up of an iron and steel industry in New Zealand C. A. Blance (*Inst. Brit. Found. Australian Branch*, 1959, **10,** 35–47) The available iron ore in New Zealand is mostly present in sands. The technological and economic problems involved in setting up an iron and steel industry are considered in the light of these ores, and New Zealand's needs.

India's increasing steel production (*Engineering*, 1960, **189,** March 18, 390–391) An account of Rourkela with notes on the layout, oxygen plant, power and water supplies, and training scheme.

Pig iron production and steel manufacture G. Bulle (*VDIZ*, 1959, **101,** Oct. 21, 1405–1409) A review (83 refs).

The TsNIIChM (Central Scientific Research Institute for Ferrous Metallurgy) session for co-ordination of the principal scientific and research tasks of ferrous metallurgy in 1960 Yu. M. Shishkin (*Stal'*, 1960, (4), 383–384) Topics discussed were ore beneficiation and sintering, improvement of pig-iron for quality steel production, OH and electric steel furnace operation, continuous casting, ferro-alloys, direct processes, rolling heat, treatment, and metal physics; also the use of tracer isotopes.

Progress in metallurgical science and technology in China in the last ten years Li Hsun (*Sci. Sinica*, 1959, **8,** Dec., 1445–1458) A general account of resources with a review of factors in the intensification of steel production, of studies in progress on new methods, on ore treatment, and on new grades of steel, including non-scaling and electrical resistance alloys (Fe–Cr–Al and Fe–Si products).

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG-IRON

Recent progress in blast furnace operation at Kamaishi Iron Works, Fuji Iron & Steel Co. Ltd I. Serita, T. Yatsuzuka, J. Sawamura, and T. Fukuda (*Tetsu to Hagane*, 1959, **45,** Sept., 867–869) Operating statistics are given, for basic and foundry iron practice, with particular reference to sizing of the burden.—K.E.J.

Automation of the revolving blast-furnace distributor N. V. Mozhayev (*Metallurg*, 1959, May, 14–15) A system controlled according to the maximum and minimum temp. is described.

Under-hearth cooling for blast-furnaces E. M. Summers (*JISI*, 1960, **195,** Aug., 405–409) [This issue].

$10 mill. blast furnace now in production at Port Kembla (*Commonwealth Eng.*, 1959, **46,** July 6, 52–53) Illustrated.

Some suggestions concerning the evaluation and determination of the specific coke consumption in the blast-furnace J. Šárek and J. Hančl (*Hutn. Listy*, 1959, **14,** (11), 928–931) Various diagrams for estimating the coke rate, given in the literature, are examined and compared with curves proposed by the authors.—P.F.

Blast control in blast furnaces B. A. Moryakin (*Izvest. AN Metallurg. i Topl.*, 1959, (3), 128–132) The causes of the occasional deviations in automatic control of the pre-determined air blast volume, due to the variations in the steam supply to the driving turbine, have been investigated. The suggestions for preventing such fluctuations are made on the assumption that they are due either to non-linear links causing oscillations or to periodic disturbances. An electrical control scheme is explained which has been tested in practice and the results obtained agree with the calculated values. So for example, if in a scheme with feed-back pulses, the fluctuations in the steam supply amount to 8 t/h, with the proposed control scheme these are cut down to 0·25 t/h which is perfectly admissible.

On the operation of the axial blower for the blast furnace A. Honda, O. Mori, and T. Nozato (*Tetsu to Hagane*, 1959, **45,** Sept., 885–887) Characteristics are given for a 4000-kW blower driven by a steam turbine. Operating results are given for periods up to 7500 h.

Study on blast furnace operation by penetration of air in the tuyere zone T. Kikuchi and T. Nozi (*Tetsu to Hagane*, 1959, **45,** Sept., 873–875) Data for raceway factors and blast penetration are given for basic and foundry iron practice.—K.E.J.

Study on the travel of the blast furnace burden with radioactive isotopes. III. On the difference of burden travel and blast volume around the blast furnace periphery A. Tominaga, T. Doinouchi, T. Yatsuzuka, and S. Komaki (*Tetsu to Hagane*, 1959, **45,** Sept., 871–873) Transit time for burden and coke were determined using ^{60}Co at different peripheral positions; the blast distribution to the tuyeres was found using Pitot tubes.—K.E.J.

On the oxygen-enriched operation of Tsurumi No.2 blast furnace H. Ikegami, T. Hasegawa, and E. Sakamoto (*Tetsu to Hagane*, 1959, **45,** Sept., 869–871) Detailed statistics are given for six periods of operation, with O_2 in the blast increasing from 0 to 2·2%. Output and coke consumption increased, and coke rate fell.—K.E.J.

The coke combustion zone in a blast furnace with oxygen-enriched blast L. V. Bardina, Yu. I. Borisov, L. Z. Khodak, and A. M. Chernyshev (*Izvest. AN Metallurg. i Topl.*, 1959, (5), 11–26) The combustion of the coke by layers was studied in a laboratory water-cooled gas generator of 300 mm dia. Inside through an opening in the grate a stationary gas extracting pipe was introduced from below with its upper rim 205 mm above the grate. This level was taken as the zero level in all experiments. Below this level a bed of coke was placed in the generator and above it a stationary coke layer 165 mm thick which, during the experiment, and with the oxygen supply above it, became gradually incandescent from the top downwards until it reached the level of the air pipe and at that moment the experiment was considered at an end. The findings show that with an O_2-enriched blast the length of the combustion zone shortens. On increasing the O_2 content, the zone of maximum CO_2 concentration is found near the place where the blast enters. The maximum temp. of the combustion zone also increases in this case.

Use of a radioactive isotope in the study of blast furnace operation H. Iijima, T. Matsumoto, S. Hayashi, and Y. Yamada (*Tetsu to Hagane*, 1959, **45,** Sept., 875–877) Results are given of tests with ^{60}Co to determine retention times of liquid metal and wear of refractories.

Desulphurization of pig iron in the blast furnace B. F. Goncharov (*Stal'*, 1960, (3), 204–207) Sulphur distribution was studied at various levels and also the influence of slag components on desulphurization. Study of the hearth was made by sampling at tuyere and slag notch levels and during tapping. It is shown that half the S is lost in the range from throat to tuyere level and the rest below the tuyeres. The desulphurization is greater above the tuyeres with ore than with sinter. The process is completed mainly below slag-notch level.

Elementary considerations on the mechanics of fluids used in blast furnaces J. Le Hoch (*Mét. Constr. Méch.*, 1960, **92,** March, 133–137) The distribution of hot air blast; its feed to nozzles; their dia. and number; and their disposition in the Cowper stove are described.

Composition of the slag-forming materials in the oxidizing zone of the blast furnace M. Ya. Ostroukhov, A. V. Rudneva, and L. M. Tsylev (*Izvest. AN Metallurg., Topl.*, 1959, (1), 37–43) Petrographic analysis shows that a proportion of the slag-forming materials in the oxidizing zone are in solid or plastic condition and consist of sintered particles of iron oxides with lime and pieces of spongy iron oxides oxidized in the air stream. Also included are drops of liquid materials penetrating in the oxidizing zone which solidify in the cool airstream from the tuyeres. The analysis of the actual mineral content shows that sintering occurs in this oxidation zone with the formation of much calcium-containing silicates (alite and larnite) and calcium ferrites ($CaO.Fe_2O_3$) but the strict localized distribution of the calcium silicates in the magnetite mass of the specimens analysed shows that immediate recrystallization of the contact surfaces of the materials in plastic conditions must take place without formation of much liquid phase and without interaction of the liquid components.

Behaviour of fluorine in blast furnace smelting. IV. Mechanism of volatilization of fluorine and its movement in a 11 m³ blast-furnace smelting Pao-Tou iron ore. V. An investigation of the slag forming processes in a 11 m³ blast-furnace smelting Pao-Tou iron ore Institute of Metallurgy and Ceramics, Academia Sinica (*Acta Met. Sin.*, 1959, June, **4,** 95–106; 107–113) IV. The fluorine content of B.F. gas depends mainly on temp., and approaches

equilibrium with lime or limestone in the charge, but this only holds when a slag phase has been formed. When liquid slag appears, fluorine in the gas decreases. The CaO/F ratio decreases as the ore moves downwards; fluorine probably volatilizes from ore as SiF_4, which reacts with H_2O to give HF and SiO_2. V. The chief constituents in the primary slag are CaO, CaF_2, and wüstite. A comparison is made between slag-forming processes with Pao-Tou and ordinary acid ores.—K.E.J.

Some problems of the heat flow through the blast furnace hearth with carbon lining Wan Shu-Ming (*Acta Met Sin.*, 1959, Sept., 206–216) The relationship of carbon linings to ceramic hearths is studied. The transitional cooling effect of carbon can only appear when it is not the innermost lining. A graphite layer under the bottom behaves as under-hearth cooling. A new 'cascade' cooling method is proposed, and improvements to the thermal model for studying temp. conditions are suggested.—K.E.J.

The effect of the thickness and the thermal conductivities of the refractory linings in the blast furnace bottom on the temperature field of the bottom Wan Shu-Ming (*Acta Met. Sin.*, 1959, Sept., 195–205) The effects of thickness and thermal conductivities of the refractories, cooling conditions, and hearth temp. on the temp. field are studied by the finite Hankel transform and presented graphically. The study includes varying conductivities in multilayer hearths, and deals with critical thicknesses and 'transitional' cooling effects.

Studies on the electric pig iron smelting of iron sand. III. IV. Behaviour of various elements in reducing smelting of iron sand. Parts I and 2 A. Murakami, K. Takai, and Y. Sato (*Tetsu to Hagane*, 1959, **45,** Sept., 880–881, 881–883) The effects of Si and Ti, S, and Mn contents on various operating parameters are elucidated.—K.E.J.

The reduction of iron ore by $N_2 + CO + H_2$ A. Isimitu, A. Sigemi, and T. Higasi (*Tetsu to Hagane*, 1959, **45,** Sept., 893–895).—K.E.J.

Effect of K_2CO_3, Na_2CO_3, Al_2O_3 and SiO_2 additions on the kinetics of the reduction of the magnetic iron oxide by carbon S. S. Lisnyak and G. I. Chufarov (*Zhurn. Fiz. Khim.*, 1959, **33,** (8), 1840–1845) Reduction by charcoal takes place at lower temp. than by graphite. It has been established that additions of Na_2CO_3 and K_2CO_3 increase the rate of the reduction reaction, K_2CO_3 considerably more than Na_2CO_3. Al_2O_3 and SiO_2 may also be present during reduction as inert diluents and have no significant effect on the reduction.

Fundamental studies on iron ore reduction. II. Reduction of granular iron ore. III. Reduction of fine iron ore T. Yagi, Y. Fuke, and A. Kondo (*Tetsu to Hagane*, 1959, **45,** Sept., 890–891, 892–893) Reduction curves are given for varying amounts of H_2 at 900°C.—K.E.J.

The reduction of copper ferrite by graphite N. M. Stafeeva, V. N. Bogoslovskii, G. I. Chufarov, and V. A. Subbotina (*Fiz. Met. Metalloved.*, 1959, **8,** (5), 740–746) The reduction of the tetragonal and cubic copper ferrites takes place in stages: $CuFe_2O_4 \rightarrow Cu + Fe_2O_3$; $Fe_2O_4 \rightarrow FeO \rightarrow Fe$. At temp. between 650 and 800° only the first stage is observed and at 900° and above all the three stages. The rate of reduction in the first and third stages passes through a maximum.

Manganese reduction rate in blast furnaces T. Torigoe and S. Yabe (*Tetsu to Hagane*, 1959, **45,** Sept., 930–932) The reduction rate is correlated with Si content, slag basicity, and slag vol. Calculated values show excellent agreement with observed figures.—K.E.J.

On the manufacture of titanium slag and pig iron from iron sand with gas reduction (fundamental test) T. Ikeno and T. Hagiwara (*Tetsu to Hagane*, 1959, **45,** Sept., 877–879) Tests with four ores (TiO_2 in the range 11·1–16·9% in the fines) gave slags containing 46·6–75·6% TiO_2. The iron contained 0·080–0·456% Ti.

HyL sponge iron process (*Chem. Eng. Min. Rev.*, 1959, **51,** Sept. 15, 48–49) A brief illustrated account.

The HyL sponge iron process J. Celada, C. K. Mader, and R. Lawrence, jun. (*Iron Steel Eng.*, 1960, **37,** Jan., 87–91) The design and performance of the Monterrey plant are described. A new 500-t/day plant is under construction, and is scheduled for completion early in 1960.

Certain properties of limestones related to their use in ferrous metallurgy O. Bolgiu and S. Fruchter (*Studii Cercet. Met.*, 1959, (4), 593–610) Various indigenous rocks have been studied. The greatest reactivity was shown by the sedimentary nonmetamorphic rocks. The greater the porosity the greater also their reactivity. The reactivity diminishes with an increase in crystal size. An increase in the MgO content results in a rapid decrease of reactivity at 1350°C. Conclusions reached show that compact sedimentary limestone with fine crystals and as little MgO as possible should be used in blast-furnaces, to reduce coke consumption.

PROPERTIES, TREATMENT AND USE OF SLAGS

The effect of magnesium on the physical properties of blast furnace slags I. I. Gul'tyai, N. L. Zhilo, G. A. Sokolov, and L. M. Tsylev (*Izvest. AN Metal. i Topl.*, 1959, (3), 20–24) The investigations aimed at establishing the optimum composition of the final blast-furnace slags with a minimum viscosity and maximum desulphurization properties under conditions obtaining at the OH shop of the Magnitogorsk Combine. Calculations of the activation energy established the effect of Mg on the mechanism of the viscous flow of slags. This energy was determined by the angle of deviation of the straight line of the logarithmic relation of this energy to the inverse temperature 1/T. Results show that the activation energy for the viscous slags with minimum viscosity vary between 26·7 and 47·5 kcal/mole of melt; the lowest energy is shown by a slag with a basicity ratio CaO/SiO_2 of 1·01. Various basicities have been investigated at various temp. and result are tabulated.

Effect of the lowest titanium oxide and of ferrous oxide on the properties of high-titanium slags I. P. Bardin, I. A. Karyazin, and V. A. Reznichenko (*Izvest. AN Metallurg. i. Topl.*, 1959, (5), 35–41) The materials used for the preparation of the synthetic slags were the chemically pure oxides Al_2O_3, CaO, MgO, TiO_2, SiO_2, ferrous oxalate ($FeC_2O_4.2H_2O$), and Ti_2O_3, synthetically prepared by the reduction of the TiO_2 with metallic Ti. The viscosity of the slags was measured with a rotating electrical viscosity meter from the readings of the current in the rotor. The findings show that, inter alia, the lowest titanium oxide affects the viscosity and the m.p. of slags depending on the ratio between the Ti_2O_3 and TiO_2 which constitutes the principal technological characteristic of titanium slags.

Fluidized bed kilning of vanadium slags N. P. Slotvinskii-Sidak and V. I. Potapov (*Stal'*, 1960, (4), 327–329) Recovery was increased 5–7% by the process with acceleration of reduction and leaching.

DIRECT PROCESSES

The electric furnace and the direct reduction of iron ore I. Luca (*Met. Constr. Masini*, 1959, (7), 556–564) A general review of the position today, and a special analysis of two new methods for the direct reduction of the ore, which are, the electric furnace with shallow bath and the hydrogen method. In the first group the article reviews the Wiberg, Lurgi, and the Madaras processes and in the second, the processes of H-iron, A. D. Little (Esso Research), the Shipley process (United States Steel), Onu, and Stelling.

Brief review of the development of some iron- and steelmaking processes L. Gasenel (*PACT*, 1959, **13,** Dec., 497–499) A brief outline is given of the Elektrokemisk and Cyclosteel processes, and of vacuum casting.

Pilot plant studies new direct reduction processes (*Iron Steel Eng.*, 1960, **37,** Feb., 149–150) A description of pilot scale trials of the Orcarb process carried out by Swindell-Dressler Corp. of Pittsburgh. Pre-reduction takes place in a rotary kiln, and the resulting material is smelted in an electric furnace.

A future method of manufacturing iron which is interesting many people today. All the details of the direct manufacture of steel by the N.F. Process T. Nakajima (*Kinzoku*, 1960, **30,** March 1, 28–32) An outline of the details of the N.F. process with a test calculation of the production cost. Three actual experiments are described and also the oxidation process.

Nu-iron: good process, but C. S. Cronan (*Chem. Eng.*, 1960, **76,** April 4, 64–65) An account of the working of the pilot plant of US Steel. The process is technically successful but at present uneconomic.

PRODUCTION OF STEEL

Steel Company of Wales: Architecture to-day No.462 constructional work at Port Talbot A. T. Edwards (*Financial Times*, 1959, Feb.4, 14–15).

Tasks of our steelworks and rolling mills in the chemical industry programme R. Schröder (*Neue Hütte*, 1959, **4,** Oct., 578–584) Measures required for the development of the steel-making programme to supply the East German chemical industry are outlined, with special reference to the production of special steels.

Effect of port dimensions on open hearth output M. M. Kotrovskii and V. G. Paramonov (*Stal'*, 1960, (2), 111–117) In 370-t OH furnaces with O_2 feed, studies on port design have been made. The effects of many variables are computed and work is continuing. Certain structural parameters are suggested.

Firing the open-hearth furnace with natural gas F. Bartu (*Rev. Mét.*, 1960, **57,** Feb., 107–116) The composition of various sources of natural gas is given, and the principles of OH with auto-carburation and addition carburation are outlined, the design of burners for natural gas/fuel oil mixtures is discussed, and also the utilization of waste-gas heat. Reference is made to the economic aspects.

Converting a 220-ton open hearth furnace to natural gas firing N. N. Dobrokhotov, I. I. Kobeza, K. S. Greben', L. D. Yupko, V. T. Garchenko, and A. L. Turubiner (*Stal'*, 1960, (1), 29–32) The gas supplied to Zaporozhstal is described and the problems of conversion of a furnace design which is discussed are explained. The gas system and its control are shown. No capital outlay was involved and the changes were carried out while the furnace was operating.

Natural gas-air burner for a top-fired preheater R. J. Leary and B. J. Mitchell (*US Bur. Mines Rep. Invest.*, 5519, 1959, pp.20) The burner was designed for heating steel scrap.

Use of natural gas in Italy for firing open-hearth furnaces C. Barbazanges and J. Gardini (*Rev. Mét.*, 1960, **57,** Feb., 95–106) The distribution, production, analysis, and consumption of natural gas from Italian sources are described, and an account is given of the technique of its use in firing Italian OH furnaces. The S content is low, and the disadvantage of low flame luminosity is overcome by admixture with fuel oil.

Firing open hearth furnaces with burners through the roof A. Labat-Camy (*Blast Furn. Steel Plant*, 1959, **47,** Sept., 965–969) Trials carried out at the de Wendel Company Forges de Gueugnon are described. It is shown that substantial savings of fuel and increased production are achieved by a suitable distribution and number of burners passing through the roof, and adjustable so that the flame remains near to the scrap as it melts down.

Determining the amount of slag in basic O.H. furnaces during the refining period A. M. Skrebtsov (*Zavods. Lab.*, 1959, (9), 1078) The method here described uses ^{32}P which has an advantage over the ^{49}Ca normally used, since P disperses uniformly in the slag in a much shorter space of time than do the calcium oxides.

Working conditions for open hearth roofs with oxygen fed into the fuel stream and the bath M. M. Dvorkind (*Stal'*, 1960, (2), 117–121) Shorter roof life is observed owing to increased furnace dust concentration and increased CO just below the roof when O_2 is used.

Production, properties and economic importance of open-hearth rimming steel P. Netter (*Neue Hütte*, 1960, **5,** Feb., 63–69) Advantages and drawbacks of killed and rimming steels are compared from the points of view of production, strength, impact properties, and economics.

Refining with oxygen in open-hearth furnaces. I. Experience in Italy E. Pachaly (*Stahl Eisen*, 1960, **80**, March 31, 441–442) The author describes his impression of a visit to the Cornigliano steel works near Genoa which has 250-t OH furnaces equipped with O_2 lances which enter vertically from the roof. The output of the furnaces is greatly increased and O_2 blowing is—in the eyes of the reviewer—justified where the pig iron proportion is high and where the price for electricity is low as it is the cost of electric power that determines the price of the O_2.—T.G.

Refining with oxygen in open-hearth furnaces. II. Experience in American works G. Henke (*Stahl Eisen*, 1960, **80**, March 31, 442–444) The author reports his impressions on the use of O_2 for refining in OH furnaces in America and discusses the reasons why such practice is unlikely to be introduced on a large scale in Germany.—T.G.

The production of high-quality and high-grade open hearth steels in furnaces fired with sulphur-containing and high-sulphur fuel oil V. P. Filatov, P. P. Semenenko, N. I. Kokarev, A. G. Kapichev, and S. F. Aleksandrov (*Stal'*, 1960, (1), 36–39) Intensive combustion converting the S-compounds to SO_2 before contact with the slag or metal, by gasification in the former gas uptakes of the ports, or somewhat less effectively by direct short-flame combustion in the furnace chamber gives 0·02–0·035%S in the steel.

Investigating the working of O.H. furnace hearths by means of radioactive isotopes I. P. Bas'yas, Un. A. Danilovich, and A. N. Lepesa (*Zavods. Lab.*, 1959, (9), 1076–1077) The intermixing of iron oxides from the ores or the molten metal charge with the working layer of the hearth is here studied by means of the radioactive ^{59}Fe. This radioactive isotope in the form of metallic iron was dissolved in HCl and then precipitated with ammonia. The solution, turbid with flakes of radioactive iron hydroxide was poured into the charging box with the ore and so introduced into the furnace. This method can readily be used to determine the interaction between hearth and charge.

The effect of the oxides of magnesium and aluminium on the reaction equilibria of manganese between carbon-saturated silicon-bearing iron and lime silica slags at 1500 C H. Schenck and F. Neumann (*Arch. Eisenh.*, 1960, **31**, Feb., 83–86) The displacement of the equilibrium state of the Mn reactions in C-saturated Si-bearing iron melts was studied at 1500·C by adding MgO or Al_2O_3 to the lime-silica slag. The results are shown in an activity diagram.—T.G.

Continuous determination of the heat-absorption of the open-hearth furnace bath V. G. Lisienko and N. I. Kokarev (*Stal'*, 1960, (1), 89–92) Radiation pyrometers sited on the uptake walls were installed and wall, air, and combustion products temp. are shown to indicate heat absorption. Cold charge furnaces melting high-alloy steels and crude oil fired were used. Expressions for heat balances are given and control of the furnace operation from the data produced is foreshadowed.

The use of forsterite checkers F. L. Volovik, P. I. Gorshtein, V. D. Zelenskii, and A. M. Povarkov (*Stal'*, 1960, (2), 125–127) With magnesite-chrome roofs and high temp. the top courses of checkers were laid with forsterite bricks, but this caused the furnace to work colder, even with a checker temp. of 1400–1420·C. This was not due to thermal conductivity but was due to design where the gases and air tended to take different paths through the system. Melting of the top courses of silica brick restricted the passages and prevented this separation but the forsterite, though it failed rapidly, did not do this. Restricting the amount of the checkers used corrected the fault.

Valve reversal in open-hearth furnaces by integrating time relay Ya. S. Pinus (*Stal'*, 1960, (1), 39–41) Instead of control of valve reversal from checker temp. or temp. difference at the bottom a system using checker temp. top is described. The relay works by the continuous integration of rate of change of temp. and is shown to keep conditions more constant and to increase output. Fuel supply can also be controlled.

An important reserve of increased labour productivity in open hearth production V. M. Sobolev and S. E. Gershgorin (*Stal'*, 1959, (10), 953–955) The process of furnace repair is analysed and the labour employed is assessed. Better organization ofcold repairs and the use of standby furnaces, doing away with the need for hot repairs is advocated.

Electric furnaces, Cleveland and sunshine (*J. Met.*, 1960, **12**, Feb., 152–155) A report of the 17th AIME Annual Electric Furnace Conference, 2–4 Dec. 1959, including brief abstracts of the papers presented.

Pilot electric smelting furnace Birlec-EFCO (Melting) Ltd (*Overseas Eng.*, 1960, **33**, March, 184–185; *Iron Coal Trades Rev.*, 1960, **180**, Feb. 19, 429–430) The melting equipment available for contract service is described.

Large electric steel furnaces P. E. Hammarlund (*Tek. Tidsk.*, 1959, **89**, Aug. 14, 721–728) The design and equipment of typical large electric arc furnaces are described and illustrated.

New electric steelmaking plant. Interesting extension at Brymbo Steel Works (*Metallurgia*, 1959, **60**, July, 3–6) A description of the construction and operation of the new O_2-electric melting shop at Brymbo.

Melting high-quality steels in the induction furnace J. Dewez, R. Berger, A. Paquot, and P. Nicolaides (*AIMME Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 169–190) The authors describe work carried out at the S.A. Usines Emile Henricot to determine the exact influence of the various phases of induction melting on steel quality. Systematic tests were made to study the effect of oxidation of the bath and the specific action of various deoxidizers, particularly with respect to inclusions. It is shown that the induction furnace can work to very narrow composition ranges and produce steel comparable in quality to arc furnace steel.

The electric smelting of iron ore H. Walde (*Electrowärme*, 1960, **18**, Feb., 32–38) The author deals with the economic potentialities of the electric low-shaft furnace for the smelting of iron ore, particularly sintered fines. The electric low-shaft furnace is well suited for the consumption of surplus electric energy from power stations. The possibility of the use of the electric low-shaft furnace in an all electric steel works is touched upon.—T.G.

Electric melting economics J. G. Coutant (*Blast Furn. Steel Plant*, 1959, **47**, 1073–1078) Estimated capital and production costs are compared for OH and electric arc furnace shops (based on 125-t furnaces and regulating and other factors making for more economic electric arc furnace operation are described.

Operational experience with a 40-t arc furnace with high secondary voltage V. A. Kamardin (*Stal'*, 1960, (2), 128–129) This follows an article by Gladkii, Ivan'ko, and Kurganov and states that further experience has not fulfilled the expectation that this furnace would not increase wear of the refractories, as, in 1957, wall life fell from 29·9 to 25·7 heats. Data comparing this with other types of arc furnace are given.

A contribution to the use of liquid pig iron for steelmaking in the electric arc furnace, with special reference to the use of high phosphorus-bearing irons R. Durrer and G. Heintze (*Roll'schen Mitt.*, 1958–60, **17–19**, (1/4), pp.44) Trials were conducted in 12- and 40-t arc furnaces using a charge consisting of about 50% liquid pig iron rich in P (up to 2%) and ore, particularly fines, as a refining agent. Dephosphorization was shifted before decarburization; even when charging high-P iron, low P contents (about 0·02%) could be obtained with one slag only. The metallurgical problems involved are discussed, power consumption and its control showed that the process is competitive with other steelmaking processes owing to the high rates of refining attained. Melting practice, energy and electrode consumption, analyses, slag practice, effect of the accompanying elements and the economics ore all reported in detail. The most important feature of the process is that by adding ore to the charge a yield of 130% based on the quantity of liquid pig iron and scrap could be obtained, this being one of the main reasons for the process being competitive (34 refs).—T.G.

Use of a direct reduced iron in the electric furnace P. U. Gummeson (*AIMME, Elect. Furn. Steel Conf. Proc.*, 1958, **16**, 244–248) The author discusses the problems and possibilities in the use of sponge iron in electric furnaces, dealing particularly with Hoeganaes reduced iron produced from high-grade magnetite concentrate. Examples are quoted of the use of the material in both are and induction furnaces.—G.F.

Use of direct reduced iron in the electric furnace R. W. Farley (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 249–252) The author discusses the use of R–N direct-reduced iron in the electric arc furnace. Its melting characteristics are shown to depend on composition, physical properties, and briquette shape.—G.F.

Electrode holders H. S. Newhall (*AIMME Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 329–334) The author describes and illustrates the electrode columns and holders used in submerged-arc furnaces manufacturing ferro-alloys and calcium carbide at the Electro Metallurgical Co., Niagara Falls.—G.F.

Distribution of electrode consumption in an electric arc furnace J. Ravenscroft and M. Preston (*Metallurgia*, 1959, **60**, Dec., 253–259) Measurements in the BISRA 10-cwt are furnace are reported and related to power, input, and furnace voltage.

The control of large electric furnaces in relation to their effect on the supply system J. W. S. Payne (*Instr. Practice*, 1959, **13**, Aug., 808–814) Problems connected with the inherent instability and non-linear nature of the arc, and methods of overcoming them, are discussed.

Compressed air v. oil circuit breakers for arc furnace switching J. E. Schrameck (*Iron Steel Eng.*, 1959, **36**, Aug., 139, 140, 142) The advantages of air jet switches are set out.

Application of immersion pyrometry in an electric furnace shop W. A. Buchovecky, F. C. Langenberg, and W. C. Kollmann (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 230–240) The authors present the results of immersion temp. measurements made in the furnace before tap, in the ladle after tap, and in the ingot hot tops immediately after teeming. They show that good control of tapping temp. leads to good control of teeming temp., and that the temp. drop from tap to teem added to the liquidus temp. gives a good indication of the correct tapping temp.—G.F.

Effect of various additions on the gas content of electric arc and induction furnace steels M. J. Boris (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 121–136) A comparison is given of the gas contents and mechanical properties of plain-carbon and low-alloy steels produced in 100, 200, and 1000 lb induction furnaces and in electric arc furnaces, and the individual effects of alloy and deoxidizing additions on gas content are noted. The absence of a slag cover is considered to be the cause of higher gas contents in induction-melted steels. No essential differences in mechanical properties could be observed.

Effects of phosphorus and sulphur content on ductility and toughness of cast low-alloy steels J. Zotos (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 274–282) An outline is given of a method of producing low sulphur and phosphorus (~0·01%) steels in basic arc furnaces. The results of mechanical testing show that the reduced contents significantly improve the ductility and impact resistance of cast low-alloy steels.—G.F.

Desulphurization of steel by injection of lime and aluminium magnesium alloy D. E. Parsons and S. L. Gertsman (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 283–298) The authors have carried out a series of tests on desulphurization of steel by the injections of a mixture of lime and Al–Mg alloy. The technique is described and results are quoted for both acid and basic electric steels. With initial sulphur contents in acid steel of up to 0·05%, about 40% reduction was obtained; with basic steel, sulphur was reduced from 0·02% to 0·008%.—G.F.

Split heat techniques G. J. Pride (*AIMME, Elect. Furn. Steel Conf. Proc.*, 1958, **16**, 313–

321) At Canadian Unitcast-Steel Ltd, small orders for low-alloy steels are met by pouring a specified amount of a standard steel into a smaller ladle and adding finely divided ferro-alloys, nickel, carbon, etc. The practice is described by reference to specific examples, and the special precautions which are necessary are discussed.—G.F.

Split heats at Texas foundries S. O. Smith (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16,** 322–323) Brief details are given of 'split-heat' practice at Texas Foundries Inc. The technique involves re-ladling into smaller ladles in which the alloying additions are made.—F.G.

Split heat techniques J. W. Tibbits (*AIMME Elect. Furn. Steel Conf. Proc.*, 1958, **16,** 324–325) Canadian Steel Foundries Ltd employs OH furnaces from 15 to 50 t, a 7½-t acid electric furnace, a 4-t basic electric furnace, and a 138 lb induction furnace. Split heats are made, involving re-ladling with alloying additions in the ladle, and also diluted heats in which a small heat of highly-alloyed steel is diluted with a lower-alloy metal from a second furnace. Brief details of the practices are given.—G.F.

Split heats at Battelle Institute N. H. Keyser (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16,** 326–327) Split heats in the research foundry at Battelle Institute are made by pouring only part of the heat at a time from the induction furnace, followed by alloy additions either to the ladle or to the metal left in the furnace. Brief details of the practices are given.—G.F.

Structure and properties of semi-killed St5ps steel A. G. Kalashnikov, M. Sh. Belogloveskii, and M. T. Bul'skii (*Stal'*, 1960, (2), 153–158) Methods of production and properties of St5ps steel in the semi-killed state are discussed. This gives a 10–12% saving as compared with fully killed steel and with the best techniques the mechanical properties are not inferior. The absence of Si is compensated by increase in Mn. However, further improvements are desirable.

Predicting minimum materials cost for stainless steels D. C. Hilty, R. W. Taylor, and R. H. Gillespie (*Blast Fur. Steel Plant*, 1959, **47,** Sept., 962–964) The results are reported of a linear programming investigation on the economics of stainless steel melting in USA, taking into account sources of raw materials, and melting variables.

Finkl and Sons develop process to degass steel in the ladle (*Iron Steel Eng.*, 1959, **36,** Sept., 192) In this process degassing of the contents of the ladle is followed by conventional teeming.

FOUNDRY PRACTICE

The mechanization of difficult operations in small casting shops V. F. Gramzin (*Lit. Proizv.*, 1959, (11), 18) The standard shaking-out machines used at the Voronezh foundry, which makes ore concentration equipment, have certain defects, among which are the great amount of dust which they produce and the clogging of the sieves when shaking out moist moulding materials. Various improvements are suggested.

Foundry mechanization and automation H. Rosenberger (*Technik*, 1959, **14,** Sept., 586–594) The aims of foundry mechanization are discussed, and examples of the mechanization of sand, shell, and permanent moulding are reviewed, together with other examples of mechanization and automation.

The B.C.I.R.A. experimental cupola installation (*Metallurgia*, 1959, **60,** July, 12–16) An account of the cupola design and equipment. It may be operated with cold or hot blast, using acid, neutral, or basic slags and with or without water-cooling. Data are given of air supply and heating, charging equipment, molten metal disposal, and instrumentation.

The B.C.I.R.A. experimental cupola installation H. J. Leyshon and R. I. Higgins (*BCIRA J.*, 1959, **7,** June, 658–666) The principal features of the plant are described and illustrated. It can be adapted for cold or hot blast melting, with or without water cooling, and with acid, neutral, or basic slags.

Experimental cupola is highly flexible H. J. Leyshon (*Foundry*, 1959, **87,** Oct., 89–91) The design and operation of the new experimental cupola constructed by BCIRA at Alvechurch are described.

Examination of the TsNIITMASh cast-iron melting installation I. O. Tsypin and M. K. Sin (*Lit. Proizv.*, 1959, (11), 31–35) Here are some details: 600 mm dia. cupola shaft, two rows of tuyeres, four in a row. Distance between rows 270 mm; closed top, two-stage recuperator. Scrubber, blower of 6700 m^3/h capacity. Experimental melts show that with a blast preheated to 450–500° the yield of the cupola may increase by 10%, often only 5–8%. The preheating of the blast resulted in burning of the linings in spite of the water cooling. The sulphur content in the iron melted with a preheated blast is between 0·05 and 0·1% which on an average is substantially lower than in the iron melted with a cold blast.

Some notes on monolithic cupola linings, their production and application S. Fransson (*Gjuteriet*, 1959, **49,** (8), 193–194, 197–198, 201–203) The composition of rammed linings for acid cupolas is discussed; important factors are grain structure and shape, low clay content (critical Al_2O_3 content is 7–10%). Some details of lining and patching procedures are given.

Carbon blocks help spur output in two giant-size cupolas (*Iron Age*, 1959, **184,** Oct. 8, 118–119) Two 11 ft dia. cupolas at Riverdale, Illinois, are carbon lined at the well zone. Diamond-saw machined carbon blocks up to nearly 3 ft in height have been used.—A.G.

Effect of coke properties on melting in the cupola W. Patterson and F. Neumann (*Giesserei*, 1960, **47,** March 24, 131–136) The behaviour of coke in the various zones of the cupola is discussed and the coke properties required are derived. The main factor appears to be specific surface of the coke involving grain size and strength. For cupola practice a coke of even quality and uniform size is of great importance for producing consistently good iron.

Topical problems of cupola melting and their solution abroad Ya. G. Polyakov (*Lit. Proizv.*, 1959, (11), 44–48) A survey of the position abroad on the question of cold or hot blast. Almost all references are Western European, British, and American.

The thermodynamics of the water-cooled hot-blast cupola K. Roesch and H. Huber (*Giesserei*, 1960, **47,** March 24, 145–148) In the melting of malleable cast iron in the hot-blast cupola, water-cooling by sprinkling the outside of the cupola is of advantage only when consistently high throughputs are required, in all other cases it is a positive disadvantage. Where it is used coke consumption is higher, the sulphur content of the iron is higher if an acid lining is employed, the iron is cooler, the losses in alloying elements—particularly in silicon—are higher, and finally the water consumption is rather high, but the life of the cupola and particularly of the lining is greatly extended.—T.G.

Hotter blast in the cupola A. B. Frenkel' (*Lit. Proizv.*, 1959, (11), 43) The Likhachev Automobile Plant last year commissioned a new plant for heating the cupola blast to 500–550° with town gas burnt in the furnace. This is fitted with six multi-nozzle burners. This new heater, which has operated for the last three months, has raised output by 25%, reduced coke consumption per t of metal from 150 kg with cold blast to 110–100 kg with hot at 400° and to 90–70 kg with blast at 500°.

Use of calcium carbide in the cupola H. Timmerbeil (*Gjuteriet*, 1959, **49,** (8), 203–204; from *Giesserei*, 1958) Coke economy, increased production and reduced S content are among the advantages to be obtained by additions of CaC_2.

Advantages of melting with calcium carbide additions in the cold-blast cupola G. Stephan (*Giessereitechn.*, 1959, **5,** June, 176–177) It is pointed out that CaC_2 can be used to increase economically the iron temperature with simultaneous coke charge reduction; this is illustrated by an example with operating data.

Continuous carbon injection J. E. Wilson and R. C. Shnay (*Mod. Castings*, 1959, **36,** Oct., 55–60) Fine particles of graphite carried by a stream of N_2 through a graphite lance immersed in metal are used to increase the carbon content of iron melted in an acid-lined cupola. Tests in two foundries established a practice which gave consistent and economic results.—A.D.H.

Importance and possibilities of weighing all materials in cupola practice H. Weinberg (*Giesserei*, 1960, **47,** March 24, 148–154) The economic importance of weighing everything that is fed into the cupola is stressed and weighing machines are described suitable for the large and the small foundry.—T.G.

Savings through cupola meters (*Iron Steel*, 1960, **33,** Feb., 78) A cupola meter, manufactured by John Thompson Instrument Co., and used for controlling blast on two cupolas, has led to uninterrupted melting without trouble with cold metal, production has increased, and coke and refractory wear have diminished.

The efficient use of coke in foundries N. K. Leeder (*Castings*, 1959, **5,** Oct., 7, 9, 11, 13, 15) Some factors in the production and handling of metallurgical coke which influence its properties in foundry use are assessed.

The economic use of fuels in foundries—oil fuels H. R. Goode (*Castings*, 1959, **5,** Oct., 33, 35, 39–41, 43–48) Properties, thermal characteristics handling, and safety aspects of oil fuel in foundry applications are discussed, and costs are analysed.

The economic use of gas in foundries W. J. Gomes (*Castings*, 1959, **5,** Oct. 17, 19, 21, 23, 25, 27) A review of burner design with notes on control and a comparison of costs with solid and liquid fuels and electric power.

The overheating of iron in the cupola A. V. Kaibichev and V. P. Chernobrovkin (*Lit. Proizv.*, 1959, (12), 22–24) A study of the effect of structure, rate of heating, and the oxidizing processes on the overheating of iron. The effect of the rate of melting and structure on the melting temperature was studied on specimens of white and grey iron of definite compositions. With an increase in the temp. of the working space in the cupola from 1380–1670° the temp. of the droplets of white iron on detachment increased by 100°. The white iron droplets overheat more intensely than those of grey iron. The effect of the oxidizing processes was studied by determining the losses occurring while the liquid iron seeps through the same bed charge with and without air blast.

The moisture content of air in cupola operation R. Radtke and W. Laqua (*Giessereitechn.*, 1959, **5,** June, 173–176) In view of contradictory results appearing in the literature on the effects of water additions to the cupola blast, trials carried out on the hot blast cupola of VEB Leopziger Eisen- und Stahlwerke are described. No great advantages followed from the addition of water, but good results were obtained with about 12–14% H_2O per kg of dry air.

The behaviour of sulphur in the cupola K. Stockkamp (*Giesserei*, 1960, **47,** March 24, 137–144) The reactions involving sulphur and taking place in the cupola are described, and the factors derived that either eliminate sulphur from the iron or prevent it from being taken up. Cupola practice including the design of the cupola is also dealt with.—T.G.

Computer for apportioning the burden of the cupola V. I. Loskutov (*Lit. Proizv.*, 1959, (11), 19–20) To obtain the suitable quality of cast iron, a computing machine has been developed which permits rapid apportioning of the corresponding burden. For this purpose it has been assumed that the most complex charge consists of nine elements: C, Si, Mn, P, S, Cr, Ti, Ni, and Cu. Cupola production chiefly uses scrap iron of different origins. The results given consist of the proportion of the different components per 100 kg of burden.

Dust removal from cold-blast cupolas E. Weber (*Giesserei*, 1960, **47,** March 24, 155–157) A simple dust catcher for cold-blast cupolas is described that removes about 70% of the total dust. The larger particles are removed by aerodynamic means, the finer dust is precipitated when the stream of gas is led through a water-mist screen. As the dust-catcher is simple in design and easy to operate and also not expensive to buy the smaller foundry is able to afford it. The running costs are about 9d/t of iron.—T.G.

Increasing the pressure of the air blast to the cupola V. M. Gorfinkel' and A. V. Chernetsov

(*Lit. Proizv.*, 1959, (11), 42–43) The normal air blast pressure in cupolas of 750–800 mm dia. fluctuates within a wide range from about 400 to 650 mm water-gauge. A cupola of 800 mm dia. working with a blast of 400 mm water-gauge at the tuyeres was reconstructed and the blast pressure increased to 650 mm with an increase of 20% in the volume of blast. The cupola yield was increased from 3·5 to 3·7 t/h. The temp. of the iron was raised from 1330–1370° to 1360–1390°. O_2 was fed periodically to the forehearth. As a result rejects for gas blisters fell from 3·3 to 2%.

The economic use of fuels in foundries — electricity K. A. Cameron (*Castings*, 1959, **5**, Oct., 29–31) Economics of the use of electricity in core baking, heating of patterns and moulds for shell mould castings, and in melting operations are reviewed.

30-ton electric melting furnace C. Hides and J. B. Machin (*Ing. e. Indust.*, 1959, **26**, Sept., 118–123) [In Spanish] Technical details of a 30-t arc furnace are supplied.—P.S.

Larger arc furnaces for melting shop and foundry W. J. Kelsey (*Brit. Steel.*, 1959, **25**, Sept., 332–335) Some design features of modern Metalectric arc melting furnaces are described, and the trend towards larger capacity furnaces is discussed.

High-frequency induction melting of steel in furnace units up to 5½ tons capacity S. J. Marmaroff (*AIMME Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 162–168) The induction melting experience of Babcock and Wilcox Co. covers furnaces up to 4-t capacity and the production of heats, by simultaneous melting, of up to 5½-t. The present installation is described and details are given of typical operating practice.—G.F.

High-frequency induction melting of steel in units up to 5½ tons capacity J. T. Gow and C. F. Hamilton (*AIMME Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 151–160) The Sandusky Foundry and Machine Co. operates a melting shop of six induction furnaces ranging from ½-t to 5½-t capacity. A description is given of the layout of the shop and of the furnace refractories and lining procedures used. Details are also given of the melting practices adopted and the method of composition control.—G.F.

Pig iron from scrap steel (*Iron Steel*, 1960, **33**, Jan., 33–34) The Birlefco 3-t arc furnace described has been installed at the foundry department of Ferranti Ltd to produce pig iron and special cast irons from scrap steel.

Electroslag remelting of ball-bearing and chrome-nickel tungsten constructional steels Yu. A. Shul'te, A. F. Tregubenko, V. F. Smolyakov, V. D. Maksimenko, V. P. Frantsev, S. A. Leibenzon, and I. A. Garevskikh (*Stal'*, 1960, (1), 45–50) Experimental work on these difficult grades of steel was carried out on 182 melts in the state as cast and after forging. The S and Si were found to alter during electroslag remelting and losses by oxidation depend greatly on slag composition. The slags also fall in S so that volatile compounds must be eliminated. Pure, uniform, and dense metal of high quality is obtained.

Furnaces and mixers heated by low-frequency induction in iron foundries A. Tagliaferri and P. L. Oliva (*Fond. Belge*, 1960, **30**, Feb., 44–51) An illustrated survey.

The forehearth induction mixer in the modernization of the iron foundry M. de Boe and A. Tagliaferri (*J. Four. Elect.*, 1959, **64**, May–June, 129–133) Examples are given of the design, operation, and results obtained from several types of low- and mains-frequency induction mixers.

Liquid metal cooling for consutrode melting D. E. Cooper and E. D. Dilling (*J. Met.*, 1959, **12**, Feb., 149–151) Experimental work on the use of Na–K cooling in the consumable electrode melting (of Ti) is described, and it is shown that the system is safe and practical for this type of furnace.

Continuous casting, Present horizontal process for cast iron bar Z. Stokowiec (*Iron Steel*, 1959, **32**, July–Aug., 375–378) Developments and operation of the horizontal continuous casting machine at Sheepbridge Alloy Casting Ltd are described.

Economic casting of ductile iron C. W. Gilchrist (*Foundry*, 1959, **87**, Oct., 84–88) Cost saving is achieved through elimination of risers, by control of metal composition and other measures which are described.

Ductile iron quality control S. F. Levy (*Mod. Castings*, 1959, **36**, Oct., 38–41) Methods of control of charge, metal composition, sand properties, and physical properties of annealed ductile iron castings are described.—A.D.H.

The effect of phosphorus on the mechanical properties of flake graphite cast iron G. N. J. Gilbert (*BCIRA J.*, 1959, **7**, June, 692–724) Up to 0·8%P increases tensile strength, associated with a decrease in eutectic cell size. This is found whether P is added to the melt as ferro-P, is present in the pig iron charge or is added immediately before pouring. Determination of elastic moduli showed no appreciable variation with P content. Brinell hardness increased about 40 points for addition of about 1%P (11 refs).

Casting in ceramic moulds V. M. Kolchinskii (*Lit. Proizv.*, 1959, (12), 2–3) Casting in ceramic moulds eliminates the disadvantages connected with casting steel and heat resistant alloys in sand moulds. Mixtures of marshallite and quartz are used as fillers with ethyl silicate as binder. Moulds are made without fusible patterns. Wooden patterns are used, thickly coated with alcohol-resistant varnish, so as not to interact with ethyl silicate. The moulds allow casting up to 1700°.

Casting blast furnace iron P. D. Kurichkin (*Lit. Proizv.*, 1959, (72), 3–4) Experiments are described on casting iron drawn from the standard 100 t carrier into a 2-t ladle. The iron had a content of Si+Mn between 5·8 and 6·7%. With a view to improving the quality, the iron was treated with Mg so that the residual Mg was between 0·017 and 0·060%. Castings produced from this iron were tested under industrial conditions, these castings consisted of nuts for the blast-furnace hoists, gears for pug mills and for liquid mixers, piston rings, scale cars for blast-furnace, charging boxes, etc. Results demonstrated the advantages of using Mg-treated iron direct from the blast-furnace.

Iron castings of high quality I. P. Nedosek (*Lit. Proizv.*, 1959, (12), 38) At the Gur'ev Metallurgical Works one cupola casts components for the mill machinery among which are couplings for the 500 mill. To obtain the required properties, a special charge is used with 25% steel scrap instead of the 10% normal proportion. As a result a fine structure is obtained with a large percentage of combined carbon. UTS is ~15 kg/mm² and in some cases reaches 20–22 kg/mm².

Foundry practice: casting ingot moulds from mixtures of first-melt foundry and conversion pig iron L. M. Cherkasov, L. A. Kolesnk, A. Ya. Gembera, and N. P. Nemykin (*Stal'*, 1960, (1), 93–95) The suggestion is made that the required composition could be attained by mixing foundry and low-Si irons in the ladle cars, thus extending the direct use of pig iron from the blast-furnace. The structure and life of the moulds is said to be improved.

How section size affects grey iron castings R. C. Bates (*Mat. Design Eng.*, 1960, **51**, Feb., 100–102) Testing bars were cast and graphite flake classification and relationships of size to microstructure, tensile strength, and hardness are shown. Correlation with machinability is sought.

On special malleable iron alloyed with boron and copper T. Shimomura and S. Mizunuma (*Imono*, 1959, **31**, Sept., 781–787) Malleable iron from duplexed cupola-reverberatory furnaces with 0·002–0·020%B or B-treated 0·25–1·54%C was examined. Grahpitization time was reduced up to 0·005%B and increased by more, while second graphitization was shortened by increase of B. Elongation increases and strength and hardness fall with addition of B alone. With B and Cu various advantages were found.

The production of spheroidal iron in air-tight ladles A. Jankowski, J. Piaskowski, and J. Rumor (*Fond. Ital.*, 1960, **9**, (1), 1–5) The increase in production of spheroidal iron makes the development of more economic processes necessary, in particular a suitable method for adding Mg to the molten iron. After outlining the methods in current use the authors describe a new method of producing spheroidal iron in air-tight ladles which has produced excellent results when applied on an industrial scale.

Effect of sulphur on the spheroidizing of graphite in cast iron with calcium M. Maruyama and H. Doi (*Imono*, 1959, **31**, Sept., 793–799) The minimum CaSi giving spheroidal graphite increases linearly with S content. The max. residual S in cast iron is 0·02% and above this spheroidal structures could not be obtained. Mn promotes spheroidization with <0·01%S.

Report of a study made by the titanium-containing cast iron subcommittee S. Morita (*Imono*, 1959, **31**, Sept., 811–821) [In Japanese]. **II, Committee of iron castings** (Oct., 1031–1044).

Effects of titanium on calcium spheroidal cast iron T. Kusakawa and S. Iijima (*Imono*, 1959, **31**, Oct., 1002–1008) Ti may have an antispheroidizing effect, it is more marked with Ca than with Mg iron. Mischmetal counteracts the antispheroidizing effect of Ca.

Development of green sand moulding of large castings at the steel foundry of Soc. Terni C. Jaccarino (*Fond. Ital.*, 1960, **9**, (12), 7–14) In US steel foundries almost all small and medium castings are produced by the green sand method, including castings weighing up to 5–6 t. This is mainly due to the extreme degree of specialization and mass production for which the green sand moulding technique is particularly suited. This article describes the process and emphasizes the difficulty of extending it to large castings which are not for mass production.

Structural variables influencing mechanical properties of high strength and cast steels M. C. Flemings, R. Green, and H. F. Taylor (*Mod. Castings*, 1959, **36**, Oct., 84–98) Thin plate castings of 4330 steel containing 0·27%C 1·1%Mn, 0·5%Si, 1·8%Ni, 0·9%Cr, 0·4%Mo were cast in moulds having different thermal properties to determine the effect of solidification rate on mechanical properties in the quenched and tempered condition. Strength properties were constant in all plates but the large variations in ductility were attributed to differences in microstructure and inclusion size and distribution. The differences in solidification conditions did not affect the macrostructure. The effect of drastic chilling on structure and properties of metal melted and cast under conditions of arc welding was investigated (34 refs).—A.D.H.

New look in steel castings W. H. Dunn (*Mod. Castings*, 1959, **36**, Oct., 30–32) Examples are given of the application of precision casting methods, vacuum treatment, and new alloys in aircraft castings.

Foundry designing for steel castings A. B. Steck (*Mod. Castings*, 1959, **36**, Oct., 103–104) Methods of controlling hot-tearing by control of pouring temperature and sand properties are described.—A.D.H.

Improving the quality of precision steel casting K. K. Prokhorenko and E. V. Verkhovtsev (*Lit. Proizv.*, 1959, (11), 10–11) Investment casting is carried out at the Ishevsk Engineering plant. The moulding mixture consists of 70:30 paraffin wax and stearin. The components are cast mainly from carbon steels. The metal is melted, the slag removed, Fe–Mn, Fe–Si, and Al (0·1%) added. The surface is often cracked, and marked with gas bubbles etc. due to the oxidation of the iron and of the deoxidizers by the atmospheric air. To cover the metal and insulate it against the atmosphere, different slags were placed on the open metal surface and the best results were obtained with a 4:1 sand/soda mixture with 1–1·5% H_3BO_3.

Improving the quality of steel castings I. F. Kolchin and V. V. Ryzhenkov (*Lit. Proizv.*, 1959, (11), 12–14) At the 'Sibtyazhmash' plant a chromite coating is used against crust formation. It is applied by spray to the surface of the mould and consists of 86/89% chromite, 2% dextrine, 9–12% crude molasses. Since 1955 waterglass mixtures have been used and the moulds are blown with CO_2 for 2–5 min.

The effect of the gas medium on crust formation on steel castings B. N. Blagov (*Lit. Proizv.*, 1959, (11), 7–10) A reducing, or a neutral atmosphere in the mould eliminates crust formation on castings. Tests were carried

out at 1550°±5°. The gases given out were analysed. The experimental castings were hollow and filled with cores of different mixtures. It was established that CO_2 is the main factor in crust formation. It is not, however, possible always to avoid crust formation by a reducing atmosphere. Increased pressure on the core must also be supplied as it slightly reduces the elimination of gases.

Some consideration on the nature of moulding sand G. Kashima (*Imono*, 1959, **31**, Sept., 806–810) [In Japanese].

Relation of permeability in the sand mould and gas pressure in the mould cavity M. Mutaguchi (*Imono*, 1959, **31**, Sept., 775–781) Pressure was measured by pouring Hg in a glass mould where it was proportional to the square of the rate of pouring. In standardized casting of iron in sand moulds the defects due to lack of permeability were removed by increasing the number of vent holes. The number required is calculated.

A test to control green sand properties in the mould J. Hermes, D. Bautista, G. J. Hansen, Kyaw Htun, R. G. Lundholm, P. H. Meyst, A. C. Mueller, W. F. Shaw, and J. R. Widmoyer (*Mod. Cast.*, 1960, **38**, Jan., 36–37) Methods of producing non-standard AFS sand specimens are described. Thus specimens with similar hardness and density to that of actual moulds can be produced. It is also shown that, when the hardnesses of mould and specimen are equal, then their densities are also equal.

Density–sand grain distribution effect on physical properties T. W. Seaton (*Mod. Cast.*, 1960, **37**, Jan., 45–48) The effects of sand distribution and grain distribution on the physical properties of sands in the mulled, clay-bonded, and tempered condition are presented.

Strength of moulding sand exposed to high temperature K. Hirakata (*Imono*, 1959, **31**, Oct., 1018–1024) Tests at 700°, 900°, and 1150°C are reported. Natural moulding sand shows a decrease of strength at the beginning of exposure but this can be prevented by addition of bentonite or by mulling. The synthetic sand showed no decrease.

Stickiness in core sand mixtures AFS Sand Division Committee 8-K (*Mod. Castings*, 1959, **36**, Oct., 68–70) The committee reviews the causes of stickiness and methods of overcoming this difficulty. The Retained Sand Test, although reliable in certain circumstances is not submitted as a satandard.—A. D. H.

How effective can a sand slinger or sand rammer be in a jobbing foundry A. S. Beech (*Engineer Foundryman*, 1959, **25**, June, 34–37) The author explains that a sand slinger is valuable in a jobbing foundry owing to its flexibility, high rate of working, evenness of fill, and care of the patterns.—D. L. C. P.

On the deterioration of foundry sands: collapse of foundry silica sands combined with bentonite S. Yabuuchi and M. Takeda (*Imono*, 1959, **31**, Sept., 769–774) Effects of temp. and grain size on test-pieces heated at 400–1300°C and air cooled were investigated. The pieces were then crushed by static load and cracks were looked for through the microscope. Collapse was considerable at the highest temp.

Patterns today and tomorrow R. Olson (*Mod. Cast.*, 1960, **37**, Jan., 42–43) A general description of new techniques, and of future needs and solutions.

Pattern engineering P. B. Croom (*Mod. Castings*, 1959, **36**, Nov., 38–41) The role of pattern construction in relation to casting design and production is discussed.—A. D. H.

Standardization of pattern equipment with epoxy resins N. Futterman (*Foundry*, 1959, **87**, Oct., 122–123) The Utility-Industrial Division of Burndy Corp., Norwalk, Conn., has cut down variations in patterns by partial standardization using epoxy resin patterns Some examples are given.

Principles of gating J. F. Wallace and E. B. Evans (*Foundry*, 1959, **87**, Oct., 74–81) Underlying principles of gating systems—their function, pouring rate, fluidity and oxide skin formation, application of hydraulic principles, and design detail are given, the subject being treated generally, without application to specific metals (23 refs).

Feeding distance of risers for grey iron castings G. K. Turnbull, H. D. Merchant, and J. F. Wallace (*Mod. Cast.*, 1960, **37**, Jan., 1–6) In this investigation rigid moulds were used, in order to avoid complication. It was shown that, so long as adequate feed metal to compensate for contraction is provided, self feeding of grey cast iron allows unlimited feeding distance for risers.

Mechanization of individual and small scale production A. I. Zorin (*Lit' Proizv.*, 1959, (12), 15–16) A machine for densifying by vibration is described. The densifying of the mixture in the box, the making of venting channels and the removal of the finished core can all be mechanized on this machine and as a result labour productivity has increased by a factor of 2–3.

Shell moulding machines M. Klemmer (*Giesserei*, 1959, **46**, Sept. 10, 517–524) Various types of shell moulds and their method of production are described, and German moulding machines are described and classified. Output and economic data are given (17 refs).

New polymer sand binder J. L. Dewey and T. J. West (*Mod. Castings*, 1959, **36**, Nov., 34–37) The properties of foundry sand to which 0·05% proprietary polymer and 0·05%Na tetrapyrophosphate was added are given. This new bond eliminates the need for organic bonds and enables lower amounts of bentonite to be used. The sand requires less milling, produces harder moulds with less ramming, and has better collapsibility.—A. D. H.

Proof of the product is in the casting J. H. Schaum (*Mod. Castings*, 1959, **36**, Nov., 34) The addition of 0·021% polymer binder to steel foundry sand has eliminated the use of a cereal binder and increased productivity.

Core box rigging for high production W. H. Miller (*Mod. Castings*, 1959, **36**, Oct., 99–102) The design of steel faced magnesium core boxes for the production of blown oil-sand cores with minimum maintenance is discussed.—A. D. H.

A new method of preparing casting moulds and mould cores M. P. Maidanov (*Stal'*, 1959, (10), 919) A method of working the CO_2 process is described using pure silica sand waterglass of dens. 1·4–1·45 and blowing with air + CO_2.

SB binder used for large iron castings N. P. Frolov (*Lit. Proizv.*, 1959, (12), 9–10) SP binder can be made on the spot from sulphite distillation liquor Type KBSh and the process of manufacture is explained. The correct formulae for the preparation of quick-hardening mixtures with this binder have been confirmed in practice. The entire process of moulding is described.

The effect of densification on the strength of the moulding mixture with SB binder N. D. Egorkina (*Lit. Proizv.*, 1959, (12), 31–32) The rapidly hardening moulding and core mixture with the SB binder was used for determining the relation between the physical and mechanical properties of the mixture and the degree of densification. Curves obtained by correctly observing all drying conditions may be used for assessing manufacturing conditions within a certain approximation. An approximate relation between compactness of the filling in moist and dry condition is characterized by the conversion coefficient 9, which may also be used for approximate calculations.

Control specimens for the mechanical properties of precision castings Yu. E. Bondarev and V. M. Sokolov (*Lit. Proizv.*, 1959, (11), 40–41) In reply to another article (*ibid.*, 1958, (9), N. M. Tuchkevich) the authors point out that tests on standard cast specimens are grossly erroneous on the mechanical properties of castings, thus, the yield-point on the new specimens proposed by them is 15–20% higher than on the standard form; ultimate breaking strength is sometimes 10% higher. Similarly relative elongation of the new specimens is 60–70% greater than of the old. The reduction of area also shows higher values as does impact breaking strength.

Fluidity of the moulding mass and densifying the moulds by compression G. M. Orlov (*Lit. Proizv.*, 1959, (11), 35–40) A theoretical consideration of mixture densifying by various means, compression, vibration, etc., and also by means of the fluidity of the moulding mixture.

Sand-blowing machine for making cores from chemically hardening mixtures A. M. Neimark (*Lit. Proizv.*, 1959, (11), 21–24) The internal cavities of castings for suction and force pumps and front and rear sockets for centrifugal pumps are made manually in relatively complicated cores of between 15 and 40 kg in weight. To mechanize the entire processes of filling the mould, densifying, turning it over, removing the finished core, etc., the VNIIPTuglemash has developed a four-position unit type LU-505 working with mixtures on the base of waterglass. A full description of the machine is supplied with diagrams and dimensions.

Casting design symposium R. Russell (*Mod. Cast.*, 1960, **37**, Jan., 38–40) A description of the above symposium held by EIMCO Corporation (American Foundry and Machine Company Division), and of the reactions of those at the symposium.

Casting steel blocks with external chills Yu. E. Geremrot and I. A. Pilipenko (*Lit. Proizv.*, 1959, (11), 42) Initially these castings were made in green sand. The castings had sand blisters and shrinkage holes. Later the process was changed to dry sand with external chills. Diagrams are given of the new mould which has resulted in appreciable savings.

Casting of one-piece flywheels N. D. Vasil'ev (*Lit. Proizv.*, 1959, (11), 14–15) The casting of a flywheel 8 m dia. weighing 51·3 t is described. The total amount of metal cast was 154·8 t to include allowances, sprues, and runners. The casting was annealed at 880–900° for 55 h in a pit furnace.

Nodular cast iron piston rings with a ferritic matrix T. Dumitrescu and P. Popescu (*Found. Trade J.*, 1959, **107**, Oct. 15, 319–323) A ferritic matrix containing spheroidal graphite and a relatively high Si content is said to better be and easier to achieve than the customary pearlitic material. The conditions of production, structures, and wear resistances are given with tables of endurance in service.

Improving the technology of casting worm gears N. N. Lisetskii (*Lit. Proizv.*, 1959, (11), 41–42) A modification of the original method of casting these wheels is described ensuring a reduction in the extra metal for mechanical finishing. The mould has been altered, also the position of sprues and the structure of the cogs. This has improved quality of production by 60–70%.

For economical pipe casting use the DeLavaud process G. Bernhardt (*Canad. Metalw.*, 1960, **23**, Jan., 24–27; Feb., 34–36) The present position of the DeLavaud process is described. The author also deals with casting practice and considers preparation of the machine, pouring, forming, and extracting.—T. G.

Calculation of thermal stresses and deformation of castings of constant cross-section (method of moving the neutral point) L. S. Konstantinov (*Lit. Proizv.*, 1959, (11), 27–31) A theoretical study of the bending of a solid body due to heat; addition of temp. fields and diagram of heat stresses; the calculation of deformation and stresses of a solid body by arbitrary temp. distribution over its cross-section; and finally applications of these mathematical formulae to concrete examples are described.

Suction casting A. A. Ridnyi (*Lit. Proizv.*, 1959, (12), 12–13) The production of sleeves by suction casting is described. The thick-walled mould, continuously water cooled, is immersed at its lower end to a depth of 15–20 mm into a bath of liquid alloy and this is sucked in by means of a vacuum. The installation and the mould are described, as well as the entire operation.

Casting intricate iron machine components in multi-sectional chill moulds G. I. Gordon (*Lit. Proizv.*, 1959, (12), 7–8) The casting is described of a connection to an ice machine of a very intricate design which was cast into a multi-sectional mould assembled before casting. Results were found to be satisfactory and rejects were reduced from 60–80% of the production to only 10–15% after hydraulic and pneumatic tests.

Centrifugally-cast steel pipe and other tubular products J. T. Gow (*Found. Trade J.*, 1959, **107**, Dec. 10, 579–585) An illustrated account of operations at Sandusky Foundry and Machine Co. (Ohio), is given. The casting

machine is described and the influence of rate of rotation, pouring temp., and rate and other factors are considered. Heat extraction from the surface requires control and pressure during solidification, composition, and machining allowances also need attention. The properties and inspection of the pipes are also outlined.

Process control in the investment shop J. E. Mitton (*Found. Trade J.*, 1959, **107,** Dec. 3, 560–563, discussion 563) Design factors, temp. and humidity control in the investing room, binder life, and dipcoat drying (with constant temp. and moisture recording), raw materials control and storage, the dipcoat, and investment procedures are outlined.

Shell moulds for castings D. Iturrioz (*Met. Elect.*, 1959, **23,** Feb., 69–78; March, 83–93; April 81–93) [In Spanish] A comprehensive review of shell moulding processes is given. It deals with sands and resins, making and closing the moulds, use of hot and cold pre-coated sands, making shell cores, control of the structure of the castings, and their quality.—P.S.

Shell cores with waterglass mixtures V. S. Burov (*Lit. Proizv.*, 1959, (12), 38) A core mixture is recommended which consists of 70% by weight of Bantishevsk sand, 30% spent moulding mixture, 7% waterglass, 1%NaOH (10% solution), and 0·5% pitch. CO_2 is blown for 3–5 min. for setting.

The effect of the shell mould on the castability and cooling of the cast iron M. D. Malegin (*Lit. Proizv.*, 1959, (11), 5–6) Two-layer shells were tested of different proportions of the same material, quartz sand and phenolic resin, the inner shell 1–2 mm and the outer shell 4–10 mm thick. The iron flow was studied on spiral castings. The moulds were filled under different cooling conditions in iron boxes and covered with iron shot or dry quartz sand. The granulation of the moulding mixture, content of phenolic resin, and densification of the packing do not change the iron flow to any appreciable extent. Finer sand and less phenolic resin in the moulding mixture slightly diminish the flow.

Investment casting in shell moulds: Part I, Silicate shells P. E. Schlinkmann (*Prec. Met. Mould.*, 1959, **17,** Nov., 66–67, 103, 104) A description of the process using the Nalcast range of silicate powders and binders.

Investment casting in shell moulds S. A. Miniea jun. (*Prec. Met. Mould.*, 1960, **18,** Jan., 40–41) A brief account of the properties of Glascast and the tolerances obtained. The mould making process is illustrated.

The wetting of the moulding materials by the molten alloy I. V. Valisovskii (*Lit. Proizv.*, 1959, (12), 32–35) A study to determine the angle of wetting for various moulding materials under different conditions is explained. A formula is given for the interfacial tension at the metal–quartz sand boundary calculated with the constant for the surface energy of the quartz sand. The wetting of the quartz sand and chrome-magnesite mixture by Armco iron is shown graphically; the wetting increases rapidly with an increase in grain size. The wetting angle of the molten alloy was determined for different moulding materials and for different gas media.

Some methods of reducing the degree of nucleation of a grey cast iron in relation to sinking under bosses on thin plates A. G. Fuller (*BCIRA J.*, 1959, **7,** June, 725–733) It is shown that sinking and nucleation are reduced by bubbling O_2 through the furnace, poling with a green pole, holding the metal at temp. or adding Ti to the melt, with or without subsequent CO_2 bubbling.

Surface defects in precision castings E.-G. Nickel (*Giesserei*, 1959, **46,** Sept. 10, 528–532) Externally-visible defects are classified, and illustrated examples are given, with a discussion of their possible causes and remedial measures.

Effect of Al and N on the occurrence of intergranular fracture in steel castings B. C. Woodfine and A. G. Quarrell (*JISI*, 1960, **195,** Aug., 409–414) [This issue].

Defects in permanent mould grey castings D. Junghans and G. Schwarz (*Giessereitechn.*, 1959, **5,** June, 179–185) Typical defects are classified, causes examined and measures for their prevention discussed, with examples.

Parashrinkage phenomena–veining, metal penetration, scabbing, hot tearing D. C. Williams (*Mod. Cast.*, 1960, **37.** Jan., 17–19) The origin of hot spots is discussed (and their location), and metal shrinkage, and the defects given above are explained.

Gases in cast iron B. Marinček and H. Feichtinger (*Giesserei*, 1959, **46,** Aug. 27, 489–497) Data of the gas contents of cast iron specimens melted in various furnaces and poured into Cu moulds are given. The gases present in cast iron and their behaviour are discussed and the determination of O_2, H_2, and N_2 by vacuum fusion is described and discussed (15 refs).

On scab formation in a semi-synthetic sand mould U. Harada and K. Nishiyama (*Imono*, 1959, **31,** Oct., 1009–1018) A table giving the results is shown.

An inertia shaking-out frame K. A. Chaikovskii (*Lit. Proizv.*, 1959, (12), 13–14) A 10-t inertia knocking-out frame is described. Its technical characteristics are: lifting capacity 10-t. 850 rpm of the loaded shaft, moment of inertia 16-t; greatest vertical acceleration 1000 cm/sec^2, greatest amplitude of vertical vibration 2 mm; 11 kW motor with 920 rpm. Sixteen springs of seven turns each. Dia. of wire 25 mm average dia. of spring 92 mm.

Study of the knock-out of sodium silicate cores hardened with carbon dioxide: application to the case of unalloyed steel castings M. Jeancolas and X. Virolle (*Fonderie*, 1959, (158), March, 133–135) The results obtained with various core mixes and size of castings are tabulated and discussed.

New method of trimming and fettling castings M. S. Kashanskii, M. A. Kremer, and S. I. Tysovskaya (*Lit. Proizv.*, 1959, (12), 8–9) A process is described for fettling by gas flame. It is explained that this operation may be divided into three stages based on the difference in the behaviour of the materials composing the oxide layer and the surface layer of the metal. Also air-arc fettling round the remains of risers and sprues is used after extended investigations and has resulted in standardizing these operations.

An examination of methods used to improve the surface finish of shell-moulded low carbon steel castings D. B. James and J. M. Middleton (*Brit. Found.*, 1959, **52,** Sept., 387–395) Additions of 2½%MnO_2, 2½%MnO_2+1½%PbO_2, 2½%$CaCO_3$, or 2½%Na_2CO_3 to resin bonded silica sand gave only a slight improvement to the surface of a simple step casting. Better results were obtained with zircon and olivine sand shells. Castings made in sodium silicate-bonded shells had satisfactory surfaces but tended to give cold laps on sections $<\frac{1}{4}$ in. and swelling on those <1 in.—A.D.H.

Controlling costs of foundry operations K. T. Rinderle (*Mod. Cast.*, 1960, **37,** Jan., 13–16) A method of cost control is described, in which the various shops, and individual operators have 'goals' expressed in terms that they understand. The development and application of the system are dealt with, and its advantages explained.

Cost control in the foundry D. A. Crooks (*Foundry Trade J.*, 1959, **107,** Nov. 5, 411–418) The application of a work study system, the introduction of controls and specimen time-and-motion study sheets are described and two cost control systems are explained.

Atmospheric dust in investment foundries C. M. Stoch (*Found. Trade J.*, 1959, **107,** Dec. 3, 553–557) The health hazard, sampling and counting and free SiO_2 determination, the observation of dust cloud movements and dust suppression are outlined (discussion 557–559).

VACUUM METALLURGY

Vacuum treatment of metals (*Foundry Trade J.*, 1960, **108,** Feb. 11, 169) A brief review of melting, refining, and degassing processes.

Low pressure metallurgy A. S. Darling (*Metallurgia*, 1959, **60,** Oct., 137–143; Nov., 193–199; Dec., 263–267) A review. The historical aspects are outlined up to 1917 with special reference to iron. (Nov., 193–199) In this section, cathodic sputtering and its laboratory and industrial applications are discussed, and the mechanism and applications of vacuum vaporateion are described. (Dec., 263–267) A review of industrial melting. The final article of the series.

Contribution to the study of the electric steelmaking furnace G. Grenier (*J. Four Elect.*, 1959, **64,** (3), 137–138) Various methods of casting under a vacuum are outlined briefly and compared, with some reference to the consumable electrode process.

Vacuum treatment of molten metals Birlec-Efco (Melting) Ltd (*Met. Treatment*, 1959, **26,** Nov., 421–422) A vacuum refining furnace for special alloys is briefly described. It is a 15-cwt mains-frequency induction-melting furnace housed in a tiltable vacuum tank. Turbulence of the metal in the vacuum vessel is induced by electro-magnetic means and cycle times are therefore reasonably short.—T.G.

Metallurgy and vacuum melting D. Iturrioz (*Met. Elect.*, 1956, **20,** Jun., 94–100; July, 54–58) [In Spanish] An account is given of the advantages of vacuum melting in metallurgy. Constructional details of six different commercially available furnaces are given.—P.S.

Some problems in vacuum arc melting with a consumable electrode D. A. Shabanov (*Izv. VUZ Chern. Met.*, 1958, Oct., 37–42) Relationship between ed and electrode dia. is examined, and metal loss in relation to electrode dia. and furnace pressure is investigated. Metal losses, changes in Cr content as a function of pressure, and variations in O_2 content during melting are tabulated for an Fe–Cr and an Fe–Mn alloy.

Low cost refining of molten metals (*Metallurgia*, 1959, **60,** Nov., 204, 211) An account of the Birlefco installation at the Birmingham works of Henry Wiggin and Co. Ltd, which is a degassing and vacuum refining unit for refining special alloys.

Vacuum heat-treating: A review of applications, techniques and furnace equipment F. J. Becket (*Met. Treatment*, 1959, **26,** Nov., 391–399) The field of application of vacuum heat-treatment including brazing and sintering, the furnaces used for these purposes, vacuum pumping systems, power supply, and temp. control are briefly described.—T.G.

Vacuum melting and casting K. G. Lewis (*Iron Steel*, 1960, **33,** Jan., 3–8; Feb., 47–52) Objects of vacuum melting, and basic methods of melting, casting, and pouring are outlined, and oxidation during tapping, teeming, and casting in air is considered (57 refs). Methods of ladle degassing and vacuum ingot casting are described, and the effects of pressure variations on the molten metal are discussed. Reference is made to the Dortmund–Hörder method of degassing in stages.

Vacuum casting R. H. Herrmann (*Foundry*, 1959, **87,** Oct., 82–83) Production of investment castings in vacuum furnaces as carried out by Haynes Stellite Co. of Union Carbide Corp. is described.

Kinetics of ladle degassing in the vacuum treatment of molten steel G. A. Sokolov and G. N. Oyks (*Izvest. VUZ Chern. Met.*, 1959, (1), 47–58; *HB*–4675; from *US Techn. Transl.*, 1959, **2,** Nov. 6, 670) A method of measuring rate of degassing is described and changes in H_2 and N_2 contents, total gas removed, effects of Fe–Si and other factors are set out from observations on three ranges of steels melted in 13-t electric furnace heats.

Deoxidation of steel in vacuum casting C. Royer (*Rev. Mét.*, 1960, **57,** Jan., 62–66) The lack of agreement between theory and results obtained in practice for O_2 removal is discussed, and it is shown, and verified on the operational scale, that deoxidation is controlled by the partial pressure of O_2.

An experiment on the effects of gases on vacuum-melted cast iron M. Kikuchi (*Imono*, 1959, **31,** Oct., 995–1002) Cast iron and steel with different C contents were vacuum melted and cast after holding in O_2, N_2, or A at atmos. pressure for 20 min. O_2 reduces C, more rapidly at higher temp., Si scarcely changed. In cast iron, the graphite structure was unchanged but free cementite and ferrite appeared. In the hypereutectoid steel range, free cementite was added to the initial graphite+pearlite structure. N_2 and A had no effect.

Vacuum degassing in ladle—a new technique in steelmaking C. W. Finkl (*Met. Prog.*, **76,** Sept., 111–114).

New vacuum degassing process A. Finkl &

Sons Co. (*Prec. Met. Mold.*, 1959, **17,** Nov., 68–69) A treatment for die steels removing H_2 in a vacuum chamber at about 3000°F and scavenging with He in 12 min. from starting the ejectors is outlined. The strengths and ductilities of the treated steels are shown. **Steel vacuum degassed in the ladle** (*Mat. Design Eng.*, 1960, **51,** Feb. 12) A brief description of the process and the properties of the treated metal.

Vacuum desulphurization of liquid iron alloys T. P. Floridis (*Trans. Met. Soc. AIME*, 1959, **215,** Oct., 870) A note. It was shown that S can be removed from liquid iron alloys by vacuum treatment, and that desulphurization is increased in the presence of Si, C, and Al.

Development in melting operations C. L. Corey (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16,** 116–119) The author outlines recent advances in the technology of melting stainless steels, and discussed probable future developments. Particular attention is given to vacuum melting and to fundamental studies on melting problems.—G.F.

Vacuum stream degassing A. L. Lehman (*AIMME, Elec. Furn. Steel Conf. Prov.*, 1958, **16,** 84–93) Two 250-t vacuum degassing units are in regular operation at Bethlehem Steel Co. The author describes the units and gives details of a typical cast.—G.F.

Effect of vacuum stream degassing on properties of forging steels M. A. Orehoski and J. N. Hornak (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16,** 68–84) At Duquesne Works of US Steel Corp., a practical method has been developed for vacuum casting commercial ingots of 7 to 220 t, at less than 1000 μ absolute pressure. The equipment and operating procedure are described, and the properties of vacuum-cast forgings are discussed. H_2 content is generally reduced to <1 ppm, and the forgings are less prone to 'flake' formation and have improved tensile properties. Ingot structure and segregation characteristics are unaffected. The extent of deoxidation is also studied.—G.F.

Degassing method is used for large or small heats (*Steel*, 1960, **146,** Jan. 11, 78–81) Two installations in USA using the Dortmund Horder Huttenunion vacuum treatment for degassing steel are described. In this treatment a dip-pipe in the bottom of the vacuum vessel penetrates the molten metal in a ladle; by raising and lowering one of the vessels a portion of the steel enters the degassing vessel and returns to the ladle and this is continued until the required effect is obtained. One of the American installations is for deoxidizing large heats of special steel and the other to effect hydrogen removal to prevent flaking in forgings. Treatment costs are estimated at $2.75 to $5 per t.—D.L.C.P.

Decomposition of nonmetallic inclusions in [liquid] steels A. F. Vishkarev and V. V. Kondakov (*Proizv. Obra. Stali Splavov*, 1958, (38), 196–208; *HB*–4598; from *US Techn. Trans.*, 1959, **2,** Sept. 18, 449) The extent and rate of decomposition of inclusions of MnO; MnO, FeO; Cr_2O_3; Cr_2O_3, FeO; and silicates and aluminates in liquid steel in a vacuum and mechanisms and factors affecting them are studied.

The manufacture of rotor shaft forgings from vacuum-cast ingots A. A. Markaryants, Yu. D. Smirnov, A. D. Men'shikov, and B. F. Emel'yanov (*Stal'*, 1960, (2), 148–152) The vacuum-casting equipment and process are described and the improved metal quality arising from degasification is noted. Ingot structures are carefully examined and the simplification of forging produced is shown. More work is in progress.

REHEATING FURNACES AND SOAKING PITS

Improvement of soaking pit heaters by means of scale models S. Chernokh, I. Repiski, and M. Michek (*Stal'*, 1960, (2), 178–181) Scale models one-tenth full size using water with suspended mustard seeds were used to improve the design and the operation of foreign-designed pits is described. Four-burner designs were shown to be 25% more efficient.

Closed circuit television in the Fairless works R. W. Sheffler (*Iron Steel Eng.*, 1959, **36,** Sept., 167–171) The use of television screens in the plant at the slab reheating furnaces, bloom and billet mills open-hearth plant (for direct observation of bath and refractory conditions), and other applications are described.

Application of the doughnut rotary furnace to production of seamless steel tubing G. S. Coleman (*Blast Furn. Steel Plant*, 1960, **48,** Feb., 179–182) The design and operation of the rotary furnaces used for heating round billets prior to piercing at the Ohio Seamless Tube Division of Copperweld Steel Co. are described.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Common controlled furnace atmospheres (*Canad. Metalw.*, 1960, **23,** Jan., 29) A data sheet consisting of two tables: Gas composition and cost data, and recommended atmospheres for various heat-treating processes.—T.G.

Flames and furnaces, Types for reheating, heat treatment and refractory firing M. W. Thring (*Iron Steel*, 1959, **32,** July–Aug., 379–384) This first extract from Professor Thring's book 'The Science of Flames and Furnaces' describes with diagrams a soaking pit and some slab and billet reheating furnaces together with other types of reheating furnace. (1960, **33,** Jan., 11–15) Forms of radiation and experimental work on luminous flame radiation are considered, with a table of emissivities and the appropriate formulae. (Feb., 53–56) The effect of a number of factors on soot formation and luminous and non-luminous emissivity in turbulent jet diffusion flames is examined quantitatively.

Computer aids in design of gale velocity furnace Midland-Ross Corp. (*Steel*, 1960, **146,** Feb. 1, 84–85) Investigations into the design of convection-type furnaces are described. High velocity gas circulation requires complicated heat transfer and design calculations which have been conveniently done on a computer. The resulting furnaces have very high convection rates and excellent temp. uniformity.

Electric industrial furnaces and atmosphere generators C. Constant (*Mech. Elect.*, 1960, **44,** Jan., 24–26) A descriptive catalogue of a comprehensive assortment of electric furnaces made by Soc. MECI (some under Leeds and Northrup licence) covering forging, hardening, tempering, annealing, case hardening, brazing, etc., in a variety of types and sizes, together with a variety of endothermic and exothermic generators.—S.H.-S.

Active and neutral protective gases for the heat treatment of metals P. Fischer (*Wire World*, 1959, **1,** Oct., 61–68) A review of the usual protective gases and their uses in bright-hardening, carburization, and carbonitriding, and the control and regulation of furnace atmospheres.

Right heat treatment keeps steel stainless C. W. Sherman (*Metalw. Prodn.*, 1960, **104,** March 16, 470–474) A review classified under martensitic types, austenitic types, precipitation-hardening grades, fabricating with heat, equipment and preparation; with a short glossary of terms.

Development and process of manufacture of prestressing steel St 150 in East Germany G. von Struve (*Neue Hütte*, 1960, **5,** Feb., 78–89) Investigation of the heat treatment and mechanical properties of wire made from St 150 steel is described, with details of the results obtained from industrial-scale heat treatment and processing of 5 mm wire.

Try compromise heat treatments for dissimilar clad metals R. C. Bertossa and A. C. Fortlage (*Iron Age*, 1959, **184,** Nov. 12, 162–164) Some methods of heat treatment are described for clad metals to be used for fabricated structures or vessels. Examples given include ferritic, martensitic, austenitic, low carbon, and precipitation hardening stainless steels used with a normal carbon steel backing.

Lowering costs in heat-treating operations (*Inco-Mond Mag.*, 1959, (11), 4–7) The use of Cronite boxes for use in salt-baths, etc., is described.

The heat treatment of steel R. Schamschula (*maschinenw. Elekt.*, 1960, **15,** March, 111–119) In this part of his review the author deals with furnaces or other means of heating for all ranges of temp., equipment for quenching and cooling temp. measurement and testing, including tests for hardenability and surface cracks.—T.G.

Hardening of steel rolls (*Engineer Foundryman*, 1959, **25,** June, 39) News is given of the first ACEC roll hardening machine to be installed in Africa. Plant includes induction heating equipment for softening and hardening the rolls, and a hot-air tempering furnace.

The hardening and heat treatment of nodular cast iron W. Patterson and D. Ammann (*Härt. Techn. Mitt.*, **13,** (1), 30–45) Chemical composition and constitution, physical properties, quench hardening and annealing, isothermal annealing, surface hardening, and the influence of alloying additions are discussed (38 refs).

Nickel alloy steels: case hardening (*Mond Nickel Co. Publication*, 1959, pp.36) The processes are described with sketches of equipment. Heat treatment and effects of grain size, section size, temp. and time are given and properties and tests are tabulated.

Automatic hardening and tempering of carbine barrel forgings (*Wild Barfield Heat-Treat. J.*, 1960, **7,** March, 8–10) A fully automatic furnace installation is described. It consists of a pusher-type hardening furnace with automatic quench and a pusher type-tempering furnace, also with quenching and degreasing plant.

Choice of burner is the key to successful flame-hardening B. J. Ward jun. (*Metalw. Prodn.*, 1960, **104,** Feb. 17, 289–292) Various types and designs of burners are considered, described, and illustrated and the method of moving the burner is indicated for a number of applications.—T.G.

How and when to use induction heating (*Steel*, 1959, **145,** Dec. 28, 82–86) An extensive review of equipment and its uses.

The importance of purity and uniformity of steels for induction or flame hardening O. Patterman (*Usine Nouv.*, 1959, **16,** Dec. 31, 37–38) After referring to a study on the subject made in 1952 (*Machine Outil Française*, 1952, **64,** March) further advances made in the uses of steel for induction and flame hardening are described and tabulated.—S.H.-S.

High frequency induction heating in mechanical engineering M. Kozak (*Czech. Heavy Ind.*, 1959, (11), 3–11) An illustrated review of equipment.

Increasing fatigue strength by soft nitriding and Sulfinuz treatment J. Müller (*Maschinenwelt Elek.*, 1959, **14,** Aug., 367–371) The influence of bath conditions, time of nitriding and heat treatment in fatigue strength is discussed with reference to the literature and the author's work, and in connection with structural steels.

Soft-nitrided gear wheels: Results of comparative tests G. Niemann and H. Rettig (*VDIZ* 1960, **102,** Feb. 21, 193–202) The authors report on tests carried out on soft-nitrided gear wheels made of various steels. They found that load-carrying capacity was greatly increased as compared with heat-treated wheels, at the same time wear was markedly reduced. Soft-nitriding has the great advantage that parts treated do not require any machining or grinding after soft-nitriding.

Atmosphere furnace automates stainless tube annealing Trent Tube Co. (Wis.) (*Steel*, 1960, **146.** Feb. 8, 82–83) An illustrated account with a general description.

Survey of special furnaces for rocket, missile and aircraft components (*Met. Treating*, 1959, **10,** Sept.–Oct., 4–7).

Modern methods of annealing wire F. Frodsham (*Wire World*, 1959, **1.** Oct., 68–73; *Wire Ind.*, 1959, **26,** Nov., 1075–1082) An illustrated account of salt bath, muffle types, and electrically heated furnaces of various kinds.

Are stress-relieved castings worth the extra cost? E. W. P. Smith (*Iron Age*, 1959, **184,** Dec. 3, 116–117) Very sensitive strain gauges have been used to measure the residual stresses in a complex casting both as cast and after a stress-relieving heat treatment. A 90% reduction in these stresses was observed.—A.G.

Bethlehem develops stress relief method (*Steel*, 1959, **145,** Nov. 2, 88) Bethlehem Steel Co. have adopted an induction heat treating

method for preparing cable for concrete reinforcement. Advantages are consistent results and clean wire surfaces.—D.L.C.P.

Radiant heating: An answer to rising annealing costs R. J. Gianotti (*Iron Age*, 1959, **184,** Nov. 5, 100–101) A description of a seven-barrel-furnace line with special burners for stainless steel tubes. It uses natural gas in service or propane during idle times.

Producing high mechanical properties in resistance alloy microwire G. P. Gvozdyk (*Stal'*, 1960, (2), 169–173) High-temp. continuous annealing would be necessary to produce the necessary ductility. Higher C content assists and a protective atmosphere less reactive than dissociated NH_3 is needed.

New wire and steel annealing furnaces for Kayser Ellison (*Wire Ind.*, 9159, **26,** April, 358–380).

Recently acquired knowledge on the processes taking place in the austenitization of steel A. Rose (*Härt. Techn. Mitt.*, 1959, **13,** (1), 46–76) Austenite formation in structural steels, i.e. in the homogeneous single-phase region after equilibrium, is described and the kinetics of the process discussed, followed by a similar discussion of the austenitization of tool steels in the polyphase region, with special carbides. Carbide dissolution is discussed in relation to time and temperature. The results are applied to some examples.

Dimensional changes occurring in whiteheart malleable iron castings during annealing C. T. Moore (*BCIRA J. Res. Dev.*, 1959, **7,** Dec., 853–862) Factors influencing dimensional changes were investigated. It is shown that these changes are associated with the amount of decarburization occurring during the anneal, and that this can vary due to the variations in metal composition encountered in normal cupola practice.—D.L.C.P.

Wire coils get slow cooling in tunnel annealing furnace Timken Roller Bearing Co. (*Steel*, 1959, **145,** Nov. 2, 82) A furnace for controlling of Ni–Mo Wire coils after rolling is described. The objects of the treatment are improved grain structure for cold heading, longer die life, improved quality.—D.L.C.P.

The effect of an intermediate anneal in a hydrogen atmosphere on the specific losses of cold-rolled transformer steel A. G. Petrenko, L. A. Kurtova, G. F. Chub, M. M. Ioffe, B. N. Popov, and R. L. Sterlin (*Stal'*. 1960, (1), 71–73) Electric furnace transformer steel, uphill cast into 11·5-t ingots after vacuum treatment in the ladle was rolled to 125–130 mm slabs and then to 2·2 mm strip. This was annealed at 850° without a protective atmos. and pickled and then reduced to 0·7 mm in two passes on a cold-rolling mill. Similar strip was annealed in H_2 continuously, and other reels in DX gas and reduced to 0·5 mm in two passes. The hydrogen treatment is recommended.

Vertical cylindrical closed quench furnace Electric Resistance Furnace Co. Ltd (*Metallurgia*, 1959, **60,** Nov., 192) A description of a new furnace for gas carburizing and carbonitriding. It is particularly suitable for dense loads of parts with internal bores, and for long components.

Oil quenched iron/carbon alloys differ D. J. Fraser (*Canad. Metalw.*, 1959, **22,** May, 30–33). The author explains why modern heat treatment and quenching oils allow improved designs for component parts. The theory of quenching is explained, and factors of practical importance are stressed. Oils can be 'designed' to give the required effects.—D.L.C.P.

On the quality and heat-treatment of spheroidal graphite cast iron with added magnesium H. Hotta and T. Saruwatari (*Imono*, 1959, **31,** Sept., 788–793) Hardness, impact, and tensile tests were made at room temp. and hardness and wear tests at temp. up to 800°C. Iron from the high frequency electric furnace was hardest with oil quenching and this was reduced with tempering temp. Wear resistance was good up to 600°C and was decreased by tempering. Iron from an oil-fired furnace was inferior and ordinary cast iron poorer still.

Heat treatment of carbine barrels (*Autom. Prog.*, 1960, **5.** March, 86–87) The tempering process as carried out at a Royal Ordnance factory is outlined.

Microstructural changes upon tempering nickel chromium white iron at 400 F J. R. Mihalisin and R. D. Schelleng (*Mod. Cast.*, 1960, **31,** Jan., 32–35) It was shown that when this white iron was tempered at 400 F, the appearance of the martensite plates under polarized light is altered. Electron micrographs show that the tempered martensite has a lamellar structure, and electron diffraction results show very good agreement with published results for ε-carbide.

The effect of prestrain and retempering on ultra-high strength steel E. T. Stephenson, M. Cohen, and B. L. Averbach (*WADC–TR–57–720 Part 1, AD–206076*, 1958, Jan. 31, pp.41; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1213).

Effect of ultrasonic vibrations on the precipitation hardening and tempering of some alloys K. M. Posodina-Alekseeva and G. I. Eskin (*Metallov. Obra. Met.*, 1956, Jan., (1), 42–45) In the ferrous example described, it is shown that ultrasonic vibrations during the tempering of quenched 0·76%C steel increases hardness, apparently by ageing; the observed ageing acceleration is due mainly to the vibrations, and not to increase in temp. from absorption of ultrasonic energy.

FORGING, STAMPING, DRAWING, AND PRESSING

Some recent developments in metal working techniques B. N. Das (*NML Tech. J.*, 1959, **1,** Nov., 34–40) The author reports his impressions in this field obtained during a tour in the United Kingdom.—T.G.

New steel plant goes modern (*Iron Age*, 1959, **184,** Nov. 26, 104–106) A description is given of the new Canadian Steel Wheel Ltd plant for the production of railway wheels. Automation plays a large part in the forging, rolling, heat treatment, and finishing processes to give a production rate of 60 wheels an hour.—A.G.

A new mechanical strain gauge for measuring forces in the forming of metals F. Faulhaber (*VDIZ*, 1960, **102,** March, 262–264) A new method is suggested for measuring the forces in forging or cold extrusion of metals based on the lateral strain shown by a steel cylinder when it is subjected to compressive forces. This method of measuring the lateral strain has two advantages: the gauge can be kept very small in height and the strain itself may be measured by any conventional means such as mechanical, optical, pneumatic, or electric.—T.G.

Progressive cold forming, Swiss machines for bolt-heading and nut-forming (*Swiss Techn.*, 1959, **2,** 60–61) A universal type machine by F. B. Hatebur of Basle, specially designed to meet the lower scale of European production, as well as two machines for the mass production of nuts from hexagon bar are described.

Manufacture of Doxford crankshafts R. Atkinson (*Trans. NE Coast Inst. Eng. Shipb.* 1960, March 25, preprint 305–338) A brief historical introduction and an account of materials, heat-treatment, shrinking, machining, and assembly.

Austenitic 'cold working' for ultra high strength D. J. Schmatz, J. C. Shyne, and V. F. Zackay (*Met. Progr.*, 1959, **76,** Sept., 66–69) A description is given of 'ausforming' by which steel with high hardenability is deformed in the temp. range corresponding to the austenitic 'bay' of the TTT diagram from which the steel is quenched to martensite. The resulting structure is discussed.

The development and present position of the manufacture of components by cold forging H. D. Feldmann (*Wire*, 1959, (43), Oct., 140–142) A general outline is given of the development and present position of cold-forming techniques, particularly the production of steel parts by cold extrusion.

Influence of various trace elements on the forgeability of steel C. Rossard and P. Blain (*Rev. Mét.*, 1959, **56,** Aug.–Sept., 175–180) Forgeability was assessed by torsion tests at 900–1200°C in argon. The influence of As, Sb, and Cu was examined. All have an unfavourable effect, particularly Sb, and Cu at high temp. The effect of As is slight.

Simplified design of anti-vibration foundations for drop hammers W. Eastwood (*Engineer*, 1960, **209,** Feb. 26, 346–349) Calculations for a Massey 15-cwt ram-type hammer are presented and the possibility of ground resonance is discussed. Damping of the system and vibrations other than vertical are also considered.

Recent progress in the field of large forgings P. G. Bastien (*Mem. Soc. Ing. Civils France*, 1960, March, 45–56) Solidification of large ingots and segregation and effects of gases, the progress in press forging and heat treatment, pyrometry, vacuum casting, and non-destructive testing are reviewed.

New system of precision forging (*Techn. Indust.*, 1959, **37,** Nov., 1653–1655) Automatic machines for forging to precise limits are described; these operate on the radial forging principle and a vertical type for producing forgings of up to 300 mm dia. and 2000 mm in length is described in detail.—P.S.

Bigger, better and sounder forgings—part II E. E. Thum (*Met. Prog.*, 1959, **76,** Sept., 121–126) Investigations are described into a technique simulating the conditions of operation of large rotors. It was shown, by bursting tests on variously notched brittle and ductile discs, that the cause of rotor failure in service was a combination of some type of sharp stress concentration and a steel with a low capacity for plastic flow under conditions of high restraint. Quantitative results obtained agreed reasonably well with results of the examination of rotors fractured in service. The composition of the 'standard' composition of conventional Ni–Mo–V rotor steels has been modified with satisfactory results.

New techniques broaden forging picture—II J. H. Jackson and H. B. Goodwin (*Met. Prog.*, 1959, **76,** Sept., 89–94) The developments reviewed are—cored forging, high-energy (for example explosive forming) and other forging developments, automatic die manufacture and new control methods.

Heated dies produce better forgings P. R. Gouwens (*Metalw. Prod.*, 1960, **104,** March 16, 467–469) Dies cast from Inconel, operating at 1600°F and resistance-heated, have been developed for forging ferrous metals.

Application of press forged castings P. Gouwens, T. Watmough, and J. Berry (*Mod. Castings*, 1959, **36,** Oct., 47–54) Significant improvements were obtained in the mechanical properties of 18/8 and precipitation hardened stainless steels by forging cast blanks at cold-working temp. The process is economic for simple shapes.—A.D.H.

Quality control of drop forgings J. D. Bunton (*Inst. Prodn. Eng. J.*, 1959, **38,** Nov., 645–655) Procedures are summarized and then discussed individually. Customer relations are included. Inclusion chart methods are one topic.

Making the FN rifle at R.S.A.F. Enfield—I. Hundred operations saved by breaching P. Trippe (*Metalw. Prodn.*, 1959, **103,** Dec. 30, 2083–2090) Fully illustrated.

Forging reactor parts Bethlehem Steel Co. (*Met. Treatment*, 1959, **26,** Nov., 423–424) The special problems associated with the production of forgings for pressure vessels in atomic reactors are briefly considered. In USA for this special purpose, a new steel has been developed of nominal composition: C 0·20, Mn 0·65, Ni 0·70, Cr 0·35, and Mo 0·60%. The steel shows good weldability.—T.G.

Roll forging set-up doubles production of small parts H. Chase (*Iron Age*, 1959, **184,** Nov. 26, 98–100) A forging shop of the Brick Division of General Motors has changed over from hammer to roll forging. The differences in the processes are discussed and the advantages of the new system are given.—A.G.

The development of a practical forgeability test; effect of friction in forging and the determination of forgeability by compression test K. N. Tong, B. B. Muvdi, and G. Sachs (*MET 369–5647T2; PB* 140334, 1956, April, pp.44; from *US* Res. Rep., 1959, **32,** Aug. 14, 238) Work on Pb is described.

Machine shapes giant missile cases Pratt & Whitney Aircraft Div. (*Steel*, 1959, **145,** Dec. 21, 92) A machine for flow-turning cast steel rings into high strength pressure vessels up to 20 ft long and 80 in. diameter is illustrated.—D.L.C.P.

How to rate your press shop C. C. Higgins (*Iron Age*, 1959, **184,** Oct. 15, 112–114) Factors affecting the efficiency of an internal press shop

are discussed with suggestions for reducing costs.—A.G.

Let batch size determine press tool quality Smith-Corona Marchant Inc. (*Metalw. Prod.*, 1960, **104.** March 9, 423–425) Three classes of blanking are made to cover the range from 15000 to several million parts. These are described.

Design of eccentric presses A. P. J. Soepnel (*Sheet Met. Ind.*, 1959, **36,** Oct., 647–652, 656) In this article the writer gives information on the design of the ram guides, clearances, and forces acting on the press during working strokes, comparisons between presses of different construction, method of fitting the tool, prevention of overloading, spring cushioned connecting rods, and friction clutches.

Production of pressed steel bath tubs (*Sheet Met. Ind.*, 1959, **36,** Oct., 631–636, 655) A description of works practice at Briggs Manufacturing Co. USA producing 100 units per h. Full details are given of press-working operations and of the finishing processes which included pickling and vitreous enamelling.

On the studies of classification of press forming and formability test for sheet metal K. Yoshida (*Rep. Inst. Phys. Chem. Res.*, 1959, **35,** May, 199–213) [In Japanese].

A survey of chipless forming W. Biederstedt and H. Meyer (*Met. Treatment*, 1959, **26,** Oct., 347–360) The processes of drop-forging, hot and cold extrusion, thread rolling, roll forging, and their applications are described with reference to equipment shown at the Hanover Technical College.

Precision bending with dies F. Strasser (*Iron Steel*, 1960, **33,** Feb., 65–68) Factors influencing accuracy are discussed with examples

Mechanics of the sheet-bending process D. W. Shaffer and E. E. Ungar (*Trans. ASME J. Appl. Mech.*, 1960, **27E,** March, 34–40) Expressions for residual stresses and deformations resulting from the bending of flat sheets are obtained.

Double-acting press feeder A. C. Gustafson (*Metalw. Prodn.*, 1960, **104,** March 23, 518–519) An electro-pneumatic press feeder interlocked to the motions of the press is described.

New line tackles variety in sheet-metal processing R. H. Eshelman (*Iron Age*, 1959, **184,** Oct. 29, 124–126) An account is given of the application of automatic handling in a stamping, forming, assembly, and welding sequence. Low-cost maintenance is claimed through the use of Cypak static controls in place of conventional relays.—A.G.

Hot sizing helps alloys meet closer tolerances F. J. Peterson (*Steel*, 1960, **146,** Jan. 25, 70–72) Advice on hot sizing PH 15–7 Mo and 17–7 PH and Rene 41 alloys is given. Hot sizing effects the removal of wrinkles and correction of curvature, flatness, and flange-angle of pre-formed parts, by applying pressure and heat in tools and fixtures.—D.L.C.P.

8ounder designs, tougher steels gage universal-die output (*Iron Age*, 1959, **184,** Oct. 29, 130–131) Through the application of a modular design system the blanking, notching, and piercing of eight totally different parts is possible with only ten different die blocks. The recommenced steel for the blocks and punches is SAE–AISI–01.—A.G.

Quick chill aids hot working Ryan Aeronautical Co. (*Iron Age*, 1959, **184,** Nov. 12, 160–161) A special stainless steel requires rapid cooling in a finishing die from 700°F to 80°F. The die developed for this purpose has holes drilled in the face and sides through which refrigerated air is blown at 60 psi. With the lower die temp. a simple lead punch can be used.—A.G.

Shock wave from sparks shapes sheet metal R. Le Grand (*Metalw. Prodn.*, 1960, **104,** Feb. 17, 299–300) High-energy electrical discharges from high-voltage capacitors create shock waves which when the discharge is made under water have sufficient force to form sheet-metal parts. The process is described briefly and parts produced by it are illustrated.—T.G.

From blank to cone in one easy step H. P. Tardif (*Steel*, 1959, **145,** Dec. 7, 152) Experiments in explosion-forming cones and hemispheres are described. Exploding a charge over a flat bank on a die results in gathering and drawing effects which give a shaped piece with some of its wall thickness greater than that of the blank. Work has been done on some stainless steels; the method may be practical for small production quantities.—D.L.C.P.

Cryogenic sizing freezes accuracy into hardenable stainless steel C. H. Jung (*Product Eng.*, 1960, **31,** Feb. 15, 61–63) A cooling process that can be applied to some kinds of steel which removes strain by shaping at transformation temp. is described.

Explosive forming goes commercial Ryan Aeronautical Co. (*Steel*, 1959, **145,** Dec. 14, 98–100) An explosive forming technique is described and applications are illustrated. Parts can be formed that would be very difficult to make by conventional methods, and close tolerances are possible, so that the process can show considerable economy in suitable applications.—D.L.C.P.

The drawing of bar steel at elevated temperatures A. Mackert (*Stahl Eisen*, 1960, **80,** March 17, 374–376) A review of American practice in the production of bar steel; by varying the working temp. the bar is obtained in a temper that does not require a heat-treatment. The economic advantages arising from this type of working are discussed with special reference to German practice and economic conditions.

Bar and tube drawing W. Pöttner and K.-H. Richter (*Demag News*, 1959, (156), 31–38) An illustrated account of the processes and equipment.

The use of automatic recorders in drawing and peeling of steel rods F. Kottmeier (*Stahl Eisen*, 1960, **80,** March 17, 337–342) The 'Centralograph' is described and the economic advantages obtained from its use reported and discussed.—T.G.

Bright drawing of steel profiles H. Preussler (*Wire*, 1959, (43), Oct., 129–134) Techniques of design and production are reviewed and illustrated by examples.

Metallurgical conditions for the production of non-ageing cold-rolled sheets for deep drawing G. Odone and V. Berruti (*Met. Ital.*, 1960, **52,** Jan. 8–14) After discussing the theory of the solubility of AlN in Al-killed mild steel the authors describe the results of experiments undertaken to establish the metallurgical structure corresponding to the most desirable non-ageing properties of cold-rolled deep-drawing sheets.

Automatic processing line draws, straightens cuts and polishes bars at high speed A. I. Nussbaum (*Wire Wire Prod.*, 1959, **34,** June, 706–707, 768) The new 'Lomatic' machine is described; this produces bars of $\frac{1}{4}$ in. to $\frac{3}{8}$ in. from coiled rod at 200 ft/min., and only requires one operator. The principal piece of equipment is the high-speed heavy-duty drawbench.—D.L.C.P.

Chipless forming of rods and tubes W. Pöttner and K.-H. Richter (*Demag Nachr.*, 1959, (156), 31–38) In spite of the title, the article refers almost entirely to the drawing of tubes. General principles are outlined, and a description is given of the process and of the drawbenches and ancillary equipment used.

A graphical method for determining the drafting of dies for wire drawing machines P. Grüner (*Wire World*, 1690, **2,** Feb., 11–18).

High speed drawing of carbon steel wire W. Holloway jun. (*Wire Ind.*, 1959, **26,** April, 353–357) The development and use of high-speed wire drawing machines for high carbon steel wire by the Speedwell Wire Co. Ltd is described. Provision was made for producing wire in the range 0·010 to 0·054 in. dia. at speeds up to 3500 ft/min. The machines were equipped so that continuous operation was possible; careful preparation of the stock was provided, lubrication was by Wyrax soap compound. Quality does not suffer from the high drawing speeds. Capital costs, floor space, and operator-time are minimized with high-speed working.

Blending traditional skills with modern scientific methods, A review of the Darlington Wire Mills Ltd (*Wire Ind.*, 1960, **27,** Jan., 61–64) The plant and various processes are briefly described.

ABC of better lubrication and cooling in steel wire drawing J. G. Wistreich (*Wire Wire Prod.*, 1959, **34,** Nov., 1486–1489, 1550–1551) An investigation by BISRA is reported into the lubrication and cooling aspects of steel wire drawing. The fundamentals of these processes were investigated, and new equipment was developed. The BISRA '3-in-1' unit employs hydrodynamic lubrication which considerably lengthens the life of the die. A new block design and internal water cooling system allow considerably increased drawing speeds to be used.—D.L.C.P.

Improving the manufacturing process in the production of needle wire V. A. Golomazov and L. A. Krasinikov (*Stal'*, 1960, (2), 173–177) In preparing wire for needles from U7A, U8A, and U10A steels an evenly distributed, fine-grain pearlite structure is required. A patenting process with benzene introduced into the protective atmosphere is described.

Measurements of temperature in wire drawing with a wire drawing die thermocouple using a hard metal ball core R. Brühl (*Wire Wire Prod.*, 1959, **34,** May, 573–575, 634–636) A method of recording the temp. of a wire-drawing die during operation is described and some results are given. A hard metal ball type core was used to act as a thermocouple with the wire. The extension bar on the core, and a sliding contact on the wire formed the cold junction. Results are given for several cases.

Capacity measurement—a method for wire mills C. Burns (*Wire Wire Prod.*, 1959, **34,** May, 582–589, 633) A method for measuring the productive capacity of a wire mill shop is described. This is to give reliable information on the output of a group of machines producing products with a variety of wire sizes and insulations. An operational survey is treated statistically.—D.L.C.P.

Applications of photoelastic coatings to measurements in wire manufacture F. Zandman and F. G. Tatnall (*Wire Wire Prod.*, 1959, **34,** Nov., 1490–1491, 1552) Uses of the 'Photostress' technique, in which strains in wire are examined by observing the behaviour of plastic coating under polarized light are explained. These include—mechanism of failure of wires under tension, determining the optimum shape of wire for springs, measuring the load on the grip head in wire drawing, use as a cable extensometer and as general strain gauges.—D.L.C.P.

Forming F. Hrazdil (*Czech Heavy Ind.*, 1959, (11), 35–47) A review of cold extrusion, thread rolling, thin-walled tube bending, radial pressing, drawing, hot forming of drills, precision forming and cross rolling, mainly non-ferrous.

Extrusion presses K. Laue (*Z. Metallk.*, 1959, **50,** Sept., 495–502) Some designs of vertical and horizontal extrusion presses are described, and extrusion techniques are reviewed, with particular reference to the production of tubes and other hollow sections.

Hydraulic vertical extrusion and tube presses W. Dohrn and F. Demming (*Z. Metallk.*, 1960, **51,** March, 135–142) Hydraulic presses ranging from 500 to 200-t total pressure and working at hydraulic pressures from 200 to 315 atm. as manufactured by Schloemann AG, Düsseldorf are described, and their special fields of application outlined. Design and construction of the presses, tool design, and the hydraulic drives are included in the description.—T.G.

Cold extrusion of steel P. G. Witteman (*Prec. Met. Mold.*, 1959, **17,** Nov., 43) Suitable steels and the mechanical properties of rounds of three AISI grades (C1010, C1040, and 5120, all fine-grain) are given.

Extrusion press makes tube bigger, more concentric Scovill Mfg. Co. (*Steel*, 1960, **146,** Feb. 8, 93) A brief account of a 2500-t vertical press giving shells for cold-drawing.

Greater industrial use of extrusions (*Prec. Met. Mold.*, 1959, **17,** Nov., 42) This is inferred from the increasing sales of hot-work die steels for which figures are given.

Tool steels and heat-treatments for steel extrusions Crucible Steel Co. of America (*Prec. Met. Mold.*, 1959, **17.** Nov., 64) A table for dies, backer blocks, liners and rams, dummy blocks, and mandrels.

Metallurgical research on cold extrusion Heintz Mfg. Co. (*PB* 135708, 1948, Sept., pp.114; from *US Res. Rep.*, 1959, **31,** March 13, 163) Normalizing, quenching, and annealing gave lower hardness and more uniform grain structure for cold extrusion. The best materials tested were A 5120, C 1010 Al-killed ingot iron

and C 1010 rimmed steel which showed the largest increases in UTS on extrusion. C, Mn, Si, or Mo all increase the force required for extrusion.

Metal extrusion, Theory and basic principles of calculation S. De Filippi (*Ing. Mec.*, 1960, **9,** Jan., 7–13) The technological importance of the extrusion process is discussed. A description is given of past experiments to determine the fundamental quantities. The author sets out the calculations needed to obtain the Sachs formula which is applied in two numerical examples.

ROLLING MILL PRACTICE

Study of modern rolling P. Boer and L. Gatti (*Eng. Min. e Met.*, 1959, **30,** Oct., 197–200) [In Portuguese] A brief account of modern rolling theory is given. Roll camber, hot and cold rolling and the reasons for ingot and plate dressing are discussed.—P.S.

Determination of the resistance to deformation and of the power and work requirements in the rolling of heavy plate from plain steels between 800 and 1200 C K. E. Cetto and B. Mähner (*Stahl Eisen*, 1960, **80,** March 31, 446–448) Mathematical, the results of the calculations are presented in graphs.—T.G.

Power requirements and pass sequence for continuous rolling mills for rolling plain steels at 900–1200 C F. Claussen (*Stahl Eisen*, 1960, **80,** March 31, 448–450) Using the results of Cetto and Mahner (see preceding abstract) the author derives the pass sequence and reduction in area and the power required. The results are presented in graphs and a sample of the calculation is included.—T.G.

Torque and power requirement in rolling P. Funke jun. (*Stahl Eisen*, 1960, **80,** April 14, 536–537) A résumé of the recent work of Geleji who derived a new relation for the torque and rolling force. This relation is based on entirely new concepts of the rolling mechanism. The reviewer, however, disputes whether the new relation will find general application as the relation contains several uncertainties.

The effects of friction in the rolling of metals M. D. Stone (*Trans. ASME, J. Basic Eng.*, 1959, **81,** Dec., 681–686) The following factors are reviewed; bite angle friction; rolling friction—forward slip; the effect of friction on rolling pressure and on minimum gauge, and the effect of rolling speed on friction.

Forces on the rolls of a four-high driven back-up mill C. W. Starling (*Sheet Met. Ind.*, 1959, **36,** Oct., 667–674) In discussing the forces on the rolls there are four possible relationships between the work and backing rolls. In general the work rolls are offset to the backing rolls and tension is applied to the strip. Equations for the calculation of roll loads are given for all four cases and the author states that his analysis is sufficiently accurate for design purposes.

The distribution of contact pressures in the rolling of metals C. W. MacGregor and R. B. Palme (*Trans. ASME, J. Basic Eng.*, 1959, **81,** Dec., 669–680) Experimental determination of contact pressure distribution in the rolling of square bars and flat strip showed the existence of a double pressure peak in the bars, near entry and near the roll centre line respectively, while only the centre line peak was present in the case of the strip.

Pressure and torque measuring instruments for rolling mills T. Dietrich (*Stahl Eisen*, 1960, **80,** March 17, 372–374) Two new devices for measuring pressure or force and torque have been developed in Sweden. They are called the 'Pressductor' and the 'Torductor' and, as the names indicate, work on the induction principle. The devices and their recording apparatus are described and some applications to the rolls and stands in rolling mills are indicated.

A comparison of rolling force and torque in cold rolling by modern theories P. Funke jun. (*Stahl Eisen*, 1960, **80.** March 17, 376–377) The author reviews modern concepts in the theory of rolling-force and torque calculation and concludes that, although great progress has been made in the last few years in this field, the computation of these forces has become so much more complicated that it is now difficult to apply it (17 refs).—T.G.

Calculation of roll force in the cold rolling of strip P. O. Strandell (*Jernkont. Ann.*, 1959, **143,** (12), 810–892) Factors influencing roll force are examined in detail, namely rolling speed, roll diameter, camber and length of roll face, strip thickness, width and yield point, friction strip-roll, roll material, pass reduction, applied forward and backward tension, rolling temperature, and rolling mill elasticity. Some empirical methods for roll force calculation are described, and results of several methods of calculation are compared with experimental data; the value of various procedures is assessed (98 refs).

Surfacing of the passes in section-mill work rolls using a milling template V. K. Kobyzev (*Stal'*, 1960, (1), 68–70) Control of hard-surfacing welding is described. The process gives a satisfactory surface free from defects and not requiring machining, and the device is simple to construct, reliable in operation, and will serve for any rotating parts however complex their profiles.

Rolls for rolling mills R. A. Villanueva (*Met. Elect.*, 1958, **22,** March, 78–80; April, 80–81, 83–85, 87, 89; May, 87–97; July, 110–112; Aug., 73–76; Oct., 88–94; Nov., 78–81; Dec., 88–98; 1959, **23,** Jan., 162–166; Feb., 92–95; March, 102–107; April, 95–99; May, 84–88) [In Spanish] A comprehensive survey of the manufacture, properties, and uses of rolls is provided. Details of each aspect of the subject are given (93 refs).—P.S.

Remedy roll grinding woes H. E. Balsiger (*Iron Age*, 1959, **184,** Oct. 29, 132–134) The basic principles involved in grinding mill rolls are given together with the procedures to be followed for commercial, semi-finish, and ultra-finish surfaces.—A.G.

Fluid film bearings for vertical rolling mill rolls L. I. Malykh and A. A. Ryazanov (*Stal'*, 1959, (10), 926–929) Tests on these bearings, whose construction is described and to which thermistor circuits for warning of overheating are attached, are reported. Considerable advantages are claimed. Modes of wear and other findings are covered in detail.

The proper dimensioning of ball and roller bearings for stands in continuous section, wire and strip mills H. Pauels and H. H. Rohde (*Stahl Eisen*, 1960, **80,** March 31, 417–428) The causes of failure of bearings for stands in continuous section, wire and strip mills were investigated on a large number of failed bearings and on stands during actual working. It was found that the main causes for bearing failure are overloading, twisting during working and in assembling the rolls, and corrosion and lack of lubrication. Advice is given on a practical and theoretical basis for the proper dimensioning of bearings and for their correct mounting and servicing.—T.G.

Pipe mill bearings lubricated by grease pressure system Jones & Laughlin Steel Corp. (*Blast furn. Steel Plant*, 1959, **47,** Oct., 1079–1080) The use of multi-purpose extreme pressure grease in the line pipe mill at the Aliquippa Works and its method of application are described.

Wire rod production at new Swedish hot rolling mill (*Wire Ind.*, 1960, **27,** Jan., 55, 64) A brief description is given of an automated hot rolling mill recently put into operation by Hafors Järnwerk, Sweden. Products range from 5 in. bars to $\frac{5}{16}$ in. dia. rod. From the wire rod line, a sample is cut and guided automatically into the inspection bay.—D.L.C.P.

Rod mill has speeds up to 6,500 feet per minute (*Blast Furn. Steel Plant*, 1959, **75,** Aug 846–847) The 23-stand, 3-strand mill which began operating in June 1959 at the Kansas City plant of Sheffield Division, Armco Steel Corporation is described.

New rod mill starts operation at Kansas City (*Iron Steel Eng.*, 1959, **36,** Sept., 198, 200) The plant described is the new 23-stand, 3-strand mill, believed to be the fastest rod mill in the world, installed at the Kansas City plant of Sheffield Division of Armco Steel Corporation.

Recent development of the rolling mill industry (*Japan's Iron & Steel Industry*, 1959, Part I, 115–122) An illustrated review of recent installations. **Sizes of ordinary rolled steels** (Part V, 225–258).

High speed rod rolling—No.1 rod mill, Cuyahoga Works E. L. Chamberlain (*Wire Wire Prod.*, 1959, **34,** Nov., 1493–1499) The operation, equipment, material flow, and problems of this rod mill are described.

Calculation of metal saving by the introduction of new standards for rolled sections G. G. Bondar' and I. V. Zdornov (*Stal'*, 1960, (1), 82–84) Calculations for lightened sections and the savings produced are presented.

Thread rolling of wood screws—III B. Kunda (*Wire Ind.*, 1959, **26,** Nov., 1095–1098) The subject of groove pitch and layout is concluded and typical output figures are given.

Production of flat rolled products in the U.S.A. V. M. Piskarev (*Stal'*, 1960, (2), 140–143).

Sheet metal grader Bren Manf. Co. (*Overseas Eng.*, 1960, **33,** April, 229) An automatic grader based on measurement of thickness tolerance, perforations, and surface condition is described.

Spot sheet flaws in fast runs Chrysler Corp. (*Iron Age*, 1959, **184,** Nov. 5, 97) In the method described, flaws in sheet are detected by an ultrasonic device which then sprays the faulty section with a dye. The system can operate at coil speeds up to 400 ft/min.—A.G.

Combined wire, sheet, strip and tube mill L. Wegmann and O. Meyer (*Demag Nachr.*, 1959, (157), 7–19) A description of the Milan installation of BREDA Siderurgica Sp.A.

Hot planetary mill R. McQ. Potter (*Ing. e Indust.*, 1959, **26,** Sept., 109–117) [In Spanish] Details of the mill installed at Habershon and Sons Ltd, are given.—P.S.

B.I.S.R.A. research progress (*Iron Steel*, 1960, **33,** Feb., 63–64) The Translator screw-down control, the BISRA gaugemaster strip thickness control system, research on combined creep and fatigue, and the development of a co-ordinated forging press and manipulator, are described.

Modern electric systems in the steel industry—new concepts in design and maintenance E. H. Browning (*Blast Furn. Steel Plant*, 9159, **47,** Dec., 1271–1282) Examples are discussed of recent advances in the control and automation of electrical systems of rolling mills and processing lines. Examples include gauge control and the use of computer controls in association with basic data accumulation systems.

Organizing a steel plant preventative maintenance programme R. F. Armitage (*Iron Steel Eng.*, 1959, **36,** Sept., 100–106) It is shown how down time due to mechanical failures on a hot strip mill was reduced by 76% by a preventive maintenance programme, the development and working of which is described in detail.

Improvement of the operation of the reeling mills of a 400 mm tube rolling plant V. Ya. Osadchii, R. M. Golubchik, S. I. Vasilenko, I. I. Zuev, A. A. Shvedchenko, and N. S. Krivalidze (*Stal'*, 1960, (2), 136–139) A discussion of the working of the plant leading to the conclusions that improved pass design of the plug and work rolls, optimum setting and a satisfactory lubricant are necessary to improve the operation of the mill.

An investigation of a 1200 mm strip mill with reels in furnaces V. I. Bairakov and V. P. Fedin (*Stal'*, 1960, (2), 130–133) It is shown that carbon steel can be rolled with some degree of overheating of the drive motors when the thinnest and widest strips are produced.

Position control in rolling mills (*Autom. Prog.*, 1959, **4,** Dec., 404–405) Techniques available in Great Britain are outlined briefly.

Planetary hot strip mills set pace in productivity (*Steel*, 1959, **145,** Nov. 30, 104–110) The Sendzimir planetary hot strip mill is described and illustrated and its advantages explained. Slabs are reduced to strip in one pass. The mill can handle a variety of materials and has 'modest' capital and space requirements, so that it is useful where small capacity or small orders are encountered. There are four full size mills in operation. The technique may find particular use in conjunction with continuous casting.—D.L.C.P.

Fully automatic cold rolling (*Sheet Met. Ind.*, 1959, **36,** Oct., 657–659, 662) A description of the control system installed on a 4-high reversing mill at J. J. Habershon and Sons Ltd. This results in higher production and efficiency maintaining gauge tolerance within much closer limits and reducing the influence on

gauge and periods of acceleration and deceleration of the mill.

New cold reduction mill, £150,000 plant opened at Bromford Lane (*Iron Steel*, 1959, **32**, July–Aug., 388) A note on the new mill installed at the works of the Bromford Iron and Steel Co. Ltd, West Bromwich.

Cold rolling research and theoretical methods of determining roll pressure in the cold rolling of strip P.O. Strandell (*Jernkont. Ann.*, 1959, **143**, (11), 725–766) The development of rolling theories from Siebel and von Karman to Orowan is described. The method for calculation of roll force proposed by Bland and Ford and based on Orowan's theory is demonstrated and graphical methods are discussed. Methods of allowing for the additional forces due to elastic deformation of the strip at the entry and exit of the arc of contact in rolling hard and thin strip are described. Methods of calculating roll deformation are compared critically, and work on the experimental determination of rolling forces along the arc of contact is described (64 refs).

Use of molybdenum sulphide as a lubricant for cold rolling of tubes I. M. Pavlov, Yu. F. Shevakin, and Yu. S. Seidaliev (*Izvest. VUZ Chern. Met.*, 1959, July, 191–193) The performance of MoS_2 in the cold drawing of small-diameter alloy steel tubes, method of depositing it on the plugs, and the effect of the coating on coefficient of friction, pressure of stock on rolls and the sticking of tubes to the plug surface are discussed.

Magnetic coolant cleaning upgrades cold rolled strip Armco Steel Corp. (*Steel*, 1959, **145**, Dec. 7, 146–149) An example is cited of how magnetic separators in the coolant system of a cold rolling mill help to produce clean steel strip and reduce maintenance troubles.

Modern pipe mill at Stockton-on-Tees South Durham Steel & Iron Co. (*Brit. Steel*, 1959, **25**, Sept., 324–326) A description of the production of welded steel pipe at the new tube mill which came into production in 1955.

Recent developments in pipe and tube mill machinery D. Hancke and R. Hartenstein (*Iron Steel Eng.*, 1959, **36**, Aug., 83–91) After pointing out that, on the whole, tube mill developments in recent years have consisted of modifications of existing plant rather than new processes, the authors discuss and describe in detail several design improvements carried out on seamless tube plant constructed by Mannesmann–Meer Inc. Special reference is made to the new stretch-reducing mill at the Lone Star plant.

Manufacture of seamless steel tubes for oil wells G. Miano (*Ing. e Indust.*, 1959, **26**, Sept., 124–130) [In Spanish] Details of the layout and equipment of the finishing line for producing drill pipe and tubing at Campana are given.—P.S.

[Rolling of seamless tubes:] Presidential address of West of Scotland Iron and Steel Institute J. W. Todd (*West. Scot. ISI Paper No.520*, pp.19). The development of the Mannesmann process is outlined.

Hydraulic jet positioning device V. I. Damrin and B. I. Zobak (*Stal'*, 1960, (2), 189–197) A scheme for control of the slide gate and leading roller of the piercing stand of a tube mill depending on a jet of water passing from one nozzle to another and converting the impulse into an electric current is described.

Ebbw Vale hot mill gets new drives (*Brit. Steel*, 1959, **25**, Sept., 336–339) A description of the re-motoring of the continuous hot strip mill at Ebbw Vale, which was completed in 20 days. Details of the drives are given.

The behaviour of large electric drive motors during the reversal of the direction of rotation W. Nürnberg (*Stahl Eisen*, 1960, **80**, April 14, 514–520) The forces and moments acting in an electric motor during reversal of direction of rotation are discussed and means are described for reducing these forces. The studies were carried out with special reference to rolling-mill drive motors.—T.G.

Directing, controlling, storing, computing: Steps in the development of drive mechanisms towards automation of rolling mills H. Opitz (*Stahl Eisen*, 1960, **80**, April 14, 503–507) The latest developments in industrial electronics are briefly discussed aiming at automation of continuous strip mills.—T.G.

Check reversing mill down time General Electric Co. (*Iron Age*, 1959, **184**, Dec. 3, 114–115) For easier maintenance GEC have produced a twin drive for heavy-duty reversing mills in which the top motor is located over the extension shaft of the lower one. Other features of this new unit are an integral design of coupling and interchangeable motors.—A.G.

MACHINERY AND SERVICES FOR IRON AND STEELWORKS

Mobile plant in heavy industry—more equipment for Steel Company of Wales (*Brit. Steel.*, 1960, **26**, Jan., 26–27) A report of a speech by N. Dolling on mechanical developments in the Steel Company of Wales, the occasion being the delivery to the Company of two giant Michigan 275A supertractors—the largest built in Europe.

EOT cranes at STELCO J. Witts and J. Bird (*Iron Steel Eng.*, 1959, **36**, Aug., 92–107) Design, construction, and maintenance of the electric overhead travelling cranes at the Steel Company of Canada works, and metallurgical control in construction and maintenance are discussed in detail, reference being made to AISE crane specifications, in particular AISE standard No.6, dealing with electric overhead travelling cranes for steel mill service.

Progress in the generation, distribution, and consumption of electric power in iron and steel works G. Lemcke (*Stahl Eisen*, 1960, **80**, April 14, 498–502) A brief description of recent developments in the iron and steel industry on the electric power side.—T.G.

Experience in the operation of special-type squirrel-cage motors for steel works cranes H. Wortberg (*Stahl Eisen*, 1960, **80**, April 14, 520–525) The advantages and the limitations of squirrel-cage motors for use in steelworks cranes are discussed. Only a special type of motor particularly developed for this purpose is able to give satisfactory results. These points are also stressed in the discussion following the presentation of the paper.—T.G.

Pole-changing squirrel-cage motors for steelworks cranes H. Enke (*Stahl Eisen*, 1960, **80**, April 14, 526–531) The modifications necessary to adopt the pole-changing squirrel-cage motor to the requirements of a steelworks crane are outlined.—T.G.

The application of switchgear to steel plants R. N. Wilson (*Blast Furn. Steel Plant*, 1959, **47**, Dec., 1291–1296) A general survey is made of modern systems, including a number of outline circuits, some illustrations of equipment, and examples of the calculation of short circuits.

Dynamic characteristics of d.c. generators for arc welding M. A. Carrer (*Cien. Tecn. Solda*, 1959, **9**, May–June, 3, 1–12).

Ultrasonic welding can improve joints in dissimilar metals (*Steel*, 1960, **146**, Feb. 8, 78–79) Applications of ultrasonic welding are described. The method may be employed where other methods are impractical, unreliable, or expensive. It gives good joints between some dissimilar metals, but with present equipment one of the members being welded must be thin. Joints are strong and are effected without 'welding' so that brittleness from intermetallic compounds is avoided. The paper is made up of abstracts from an American Welding Soc. Conference.—D.L.C.P.

Electromoulding technique trims cost of parts General Motors Corp. (*Steel*, 1959, **145**, Dec. 14, 96–97) A welding technique called electromoulding is described, in which parts are hot-formed and welded to shape.—D.L.C.P.

Solved: Mach 3 manufacturing problems North American Aviation Inc. (*Steel*, 1960, **146**, Jan. 18, 86–88) Metalworking problems which have been soved to make possible the construction of the B–70 Valkyrie 2000 mph bomber are described. These include: handling, machining, and welding of the 0·006 in. steel skin; working new high duty metals; brazing of steel honeycomb using a luminous wall furnace, salt bath, electric blanket, or die quench methods.—D.L.C.P.

Iron–aluminium alloy systems part 12, Welding of iron–aluminium alloys R. Brooks and A. Volio (*WADC–TR–57–298* (Part 12); *AD–207908*, 1958, Nov., pp.39; from *Nucl. Sci. Abs.*, 1959, **13**, May 13, 1213) Fusion, resistance, spot, and stitch welding methods were developed. Coated Fe–Al electrodes for manual arc welding to deposit metal with 7–9%Al were produced and brazing problems were studied.

Joining of zirconium to stainless steel: final report J. B. McAndrew and H. Schwartzbart (*AECU–3835*, 1957, Jan., pp.127; from *US Res. Rep.*, 1959, **31**, April 17, 272: no abstract).

Join thick steel sections with electroslag welding Arcos Corp. (*Iron Age*, 1959, **184**, Dec. 17, 170–172) Electroslag welding involves using the heat generated by passing a current through a pool of molten slag to melt both filler metal and adjacent parent metal. The Vertomatic welding technique based on this principle is described and its advantages are cited.—A.G.

Ultrasonic improves quality of heat-exchanger tubing Babcock & Wilcox Co. (*Iron Age*, 1959, **184**, Nov. 19, 122–123) Methods adopted to improve the quality of welded steel heat exchanger tubing are described. These include trimming the strip just before welding, automatic removal of weld flashes, continuous ultrasonic inspection, and a continuous normalizing heat treatment.—A.G.

Explosive welding is on the way American Potash & Chem. Corp. (*Steel*, 1959, **145**, Nov. 2, 90–91) Present knowledge on the principles and practice of explosive welding is surveyed. The method has given promising results in joining a variety of metals, and may soon be commercially feasible. It may be used in conjunction with explosive forming.—D.L.C.P.

New type electrode. Shanghai–57–1 for welding of constructional steels Weng-ling Wei (*Svar. Proizv.*, 1958, Jan., 30–31) Manufacture and properties of this Chinese-developed coated electrode, giving high mechanical properties of weld and weld metal, are described.

The 'O.K43' electrode S. Rydenhag (*ESAB. Rev.*, 1959, (1), 10–12) [In French] An account of the uses of a rutile-coated welding rod.

Recent developments in weld electrodes Part 2 W. P. Campbell (*Canad. Metalw.*, 1959, **22**, April, 58–63) Several advances in electrode design and employment are described. Iron powder electrodes which allow high deposition rates are on the market. Advantage is being taken of the ability of suitable coating compositions to control the alloy composition of the weld deposit. Examples are given of electrodes that yield deposits which can be heat-treated to give tensile strengths up to 300000 psi. Control of moisture content of electrodes is becoming increasingly important.—D.L.C.P.

Development, properties and usability of low-hydrogen electrodes D. C. Smith (*Weld. J.*, 1959, **38**, Sept., 377s–392s) The history of low-H_2 electrodes, and the reasons for their development are described. The metallurgy and properties of these electrodes are summarized, and the effects of H_2 and moisture on welds, and the ways of avoiding their presence, are discussed. It is concluded that low-H_2 electrodes are superior to all other types, whatever the job, but that it is not likely that they will be generally used except for difficult-to-weld, ultra-high-strength, and free machining steels. Operators require training in the use of these electrodes, particularly in the 'vertical-up' position.

Incorporation of aqueous solutions of potassium dichromate into electrode coatings N. A. Gorpenyuk and A. V. Pechatnov (*Svar. Proizv.*, 1958, Jan., 37) The use of the solution is found to be superior to use of the solid chromate for hard-facing steel parts. The performance of the electrodes in two years of use is described.

Weldability of 3½%Ni–Cr–Mo steel C. L. M. Cottrell and K. Winterton (*Brit. Weld. J.*, 1958, **5**, Oct., 481–484) Results of continuous cooling transformation diagrams suggest that a small amount of residual H_2 (0·35 ml/100 g) is likely to have an adverse effect on the hard-zone crack resistance of a high-tensile steel. Degassing the steel was found to improve the dead-load rupture stress.

Influence of local heating on fatigue behaviour of welded specimens I. R. Gurney and

L. N. Trepka (*Brit. Weld. J.*, 1959, **6**, Oct., 491–497) Exploratory tests on mild steel specimens are reported. It is shown that local spot heating of specimens having clearly defined probable points of failure, e.g. containing discontinuous longitudinal welds, increases their fatigue strength by ~150% when the specimen is subjected to pulsating tension loading. This may be due to induced residual stresses.

Standards of welding for airframe and aircraft engine parts F. J. Tranter (*Brit. Weld. J.*, 1959, **6**, Oct., 443–455) Inspection and test methods are given which comply with those laid down by the Aeronautical Inspection Directorate of the Air Registration Board. Competency tests for operators are regarded to be of special importance. Weld defects are illustrated.—U. E.

Non-destructive testing in welding W. E. Schall (*SS News*, 1959, Dec., 10–16, reprinted from the *BEAMA Journal*) Principles, apparatus, and methods are reviewed.

On the mechanical characteristics and weldability of metallic alloys for the construction of nuclear plant F. Bragoni (*Calore*, 1959, **30**, Dec 666–680) After examining various alloys the writer concludes that Ni and Ni alloys have the best qualities for use in the construction of nuclear plant despite the high operating temp. Pressure containers and the principal parts of auxiliary equipment are now all welded using stainless steels with high Ni contents.

The strength of metal-to-metal bonds F. Eichhorn and W. Braig (*Materialprüf.*, 1960, **2**, March, 79–87) Theory and experiment show that adhesive-bonded lap joints in metals have a high load-carrying capacity if the deformation of the metals at the joints is very small, but the adhesive should show a high plastic deformation in tensile stressing and a high elastic deformability under pulsating loads.—T. G.

Weldable low-alloy steels give high strength, low weight H. Thomasson (*Canad. Metalw.*, 1959, **22**, Dec., 25, 27) Welding practice for low-alloy steels is considered for cases in which strength and weight are important factors.

Quality control methods E. Fuchs and W. Ashworth (*Brit. Weld. J.*, 1959, **6**, Oct., 429–438) Quality control methods are discussed for the purchase of welded equipment. Before manufacture the design, materials, and techniques of construction should be agreed between supplier and purchaser. Inspection procedures are described. Examples are given of a number of service failures.—U. E.

Some aspects of quality control in fusion-welded products B. K. Barber (*Brit. Weld. J.*, 1959, **6**, Oct., 480–487) Various inspection and test methods are given for ensuring good quality weldments. Emphasis is laid on the supervision and training of operators. Design features are discussed with illustrations.—U. E.

Shunters for the Steel Company of Wales Brush Traction Ltd (*Oil Engine*, 1960, **27**, Mid-Jan., 247) The new 500 hp diesel shunting locomotives are described.

How to get more for your maintenance dollar (*Iron Age*, 1959, **184**, Dec. 10, 93–108) Various aspects of the problem of maintenance are considered. Stress is placed on the need for planned maintenance under adequate supervision, the keeping of performance records, and the design of installations for easy greasing and access.—A. G.

Materials stimulus for lowering costs of railway freight handling in metallurgical establishments B. B. Rozin and R. S. Geifman (*Stal'*, 1960, (2), 165–168) The introduction of an incentive wage system is described.

Power dispatching in a steel plant C. L. Eichenberg (*Iron Steel Eng.*, 1959, **36**, Sept., 133–139) The advantages of a centralized control and direction of the generating, transmitting, and consuming units in a steel plant are discussed in relation to the application of the system to the Bethlehem plant of Bethlehem Steel Co.

WELDING AND FLAME CUTTING

New development in arc welding processes and their application P. Shaw and D. B. Tait (*Brit. Weld. J.*, 1958, **5**, Oct., 454–464) Topics discussed are—improvements in manual welding speeds by the use of iron-powder electrodes, and their limitations, and developments in semi- and fully-automatic welding processes. Several applications of automatic welding are described in the second part of the paper.

Welding of some creep-resisting steels J. F. Mercer (*Brit. Weld. J.*, 1958, **5**, Oct., 472–474) The development is described of alloy electrodes for some steels to be used for welded fabrications intended for steam raising equipment to withstand elevated temp. and pressures. Three groups of steels are considered: ½%Mo, 1–1¼%Cr, and ½%Mo, and 2¼%Cr–1%Mo (14 refs).

Some aspects of cast iron welding C. Pfeiffer and S. Lundin (*ESAB Rev.*, 1959, (1), 1–9) [In French] Properties and structures of white, grey, and nodular irons and general conditions of welding and choice of electrode are discussed. (2–3, 5–13) Nodular iron in various heat-treated conditions was welded with Ni–Fe electrodes and the results of the tests are shown. Tensile and bend tests are tabulated.

Welding of nodular cast iron C. Pfeiffer (*Gjuteriet*, 1959, **49**, (9), 225–226, 229–230, 233–234, 237–238, 241–242, 245) General problems connected with the welding of cast iron and choice of electrodes are discussed, and an account is given of the low-temp. welding of nodular cast iron at 250°C, followed by subsequent heating to 920°C and free cooling.

Cheaper way to weld tubes H. Thielsch (*Prod. Eng.*, 1959, **30**, Dec. 21, 54–55) A process using a thin insert ring in the root of the weld is described. The advantages are more rapid welding and fewer defects.

Automatic vertical welding J. A. Lucey and D. D. Smout (*Brit. Weld. J.*, 1959, **6**, Sept., 396–405) A new arc welding process has been developed in the USSR from the submerged arc welding method. The arc is initiated on some form of run-on plate and the current then passes from the tip of the electrode to the surface of the molten weld pool through molten slag. The pool is contained between the faces of the joint by damming blocks traversed vertically up to the joint together with the wire feeder as metal is deposited. The equipment is described with illustrations. The weld quality is said to be very good and no angular distortion is observed.

Some guidance on welding dissimilar alloy steels M. C. T. Bystram (*Brit. Weld. J.*, 1958, **5**, Oct., 475–480) Welding of low-alloy ferritic steels, stainless steels, and high-Ni alloys in various combinations and at low, room, and elevated temp. is discussed.

A practical approach to resistance welding K. H. McDowell and D. A. Weeks (*Brit. Weld. J.*, 1959, **6**, Sept., 381–395) Various resistance welding machines and processes used in the gas turbine industry are described. The physical characteristics of the materials to be welded are given. The metallurgical difficulties which arise when spot-welding light alloys, steels, stainless steel, nickel base alloys and Ti, are discussed with illustrations. A brief survey of testing techniques is added.

New process of assembly of steel tubes H. Goldstein (*Gén. Chim.*, 1959, **82**, Dec., 165–169) A process for end-to-end welding keeping a continuous internal tar coating is described. It also simplifies cathodic protection outside.

Metal treatment—flame welding D. Flament (*Met. Constr. Mech.*, 1960, **32**, Jan., 35–41) The process and its uses and effects on different classes of steel are described and tabulated.

Electroslag welding in action P. Trippe (*Metalw. Prodn.*, 1959, **103**, Dec. 18, 2050–2052) Mostly illustrations of the machines employed, with captions on their operation.

The new welding processes J. C. Merriam (*Met. Design Eng.*, 1960, **51**, Jan., 105–119) A comprehensive report on electron beam, ultrasonic, and 11 other welding processes, some still in the experimental stage, is presented and illustrated.—S. H.-S.

Welding by shielded arc in an atmosphere of CO_2 J. C. Rodriguez (*Met. y Electr.*, 1960, **24**, Jan., 212–216) The operation and its risks are illustrated and described.—S. H.-S.

Welding problems in the chemical industry H. Gerbeaux (*Rev. Univ. Mines*, 1960, **16**, Feb., 121–130) Welded structures of heavy materials are discussed; some details of fabrication, and the advantages of manual and automatic welding are compared. The range of materials available is discussed.

Welding of type 347 stainless steel: Minutes of the 7th Annual atomic energy commission welding conference, held in Chicago, Illinois. Nov. 1957 V. N. Krivobok (*TID*–7562, pp.13; from *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1211) High-temp. tensile tests, stress-rupture tests, and hot ductility tests, also impact toughness after ageing at 800 and 900°F, were carried out. Cracking tests with the 16–8–2 electrode are also reported.

Welding of stainless steel which contains more than one per cent boron: Minutes of the 7th annual atomic energy commission welding conference, held at Chicago, Ill., Nov. 1957 N. Balai (*TID*–7562, pp.17; *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1211) Properties of high-B cast and wrought stainless steels are given and their welding is discussed. Corrosion and irradiation tests data are given.

Joining of zircaloy to austenitic stainless steel: Minutes of 7th annual atomic energy commission welding conference. Chicago, Ill., Nov. 1957 K. H. Koopman (*TID*–7562, pp.25; from *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1211) Studies of welding, brazing, cladding, and diffusion-bonding are reported. Joints brazed with pure Ag may find application.

Selected examples of welding construction in the fabrication of equipment for the chemical industry R. Söhngen (*Cien. Tecn. Solda*, 1959, **9**, May–June, 8/8366/1–9) The special problems associated with the welding of fabricated equipment for this purpose are discussed, including the importance of protection against corrosion, and several examples are described.

Spot-welding of a 12%Cr martensitic stainless steel E. F. Nippes, W. F. Savage, L. C. Ianniello, and W. A. Owczarski (*Weld. J.*, 1959, **38**, Sept., 360s–366s) The optimum spot welding conditions were determined for 0·007, 0·015, and 0·022 in. stainless steel sheet with 0·2%C, 12%Cr and with additions of 1%Mo and 0·3%V.

Application of the automatic submerged arc welding process in assembly shops O. Grossi (*Ing. Mec.*, 1960, **9**, Jan., 19–29) The use of the automatic submerged arc welding process, although comparatively new, is becoming increasingly widespread. In this article the author reviews the industrial applications of the process outside the normal scope of the workshop, giving examples.

The useful fields of application of automatic welding K. H. Mathias (*Schweissen Schneiden*, 1960, **12**, March 102–111) The economic and quality advantages offered by the application of automatic welding methods as compared with manual welding are outlined and guides are presented for aiding in the decision of which automatic welding process is the best for a specific case (15 refs).—T. G.

Welding and cutting with electron beams K. H. Steigerwald (*Schweissen Schneiden*, 1960, **12**, 12 March, 89–95) The principles of electron-beam welding and cutting are explained, apparatus for welding, cutting and milling are described and some details of cutting and welding techniques given. Electron beams are particularly suited for the micromachining of steel as is shown on a stainless steel spinneret (0·5 mm thick) into which a slot 0·8 mm long and 0·05 mm wide was accurately cut. The main advantage of electron-beam welding is that as a result of the high energy concentration very little total energy is required thus restricting the heat-affected zone to a fraction of that occurring in any other welding process, for instance, the cross-section of the electron beam may be as small as 10^{-7} cm² but the specific energy may reach the high value of 5×10^{-9} W/cm³. As electrons penetrate sheet metal quite easily, welding can be done at places that are inaccessible to any other welding method.—T. G.

Microcracking of welds in mild steel: A review J. Sebille (*Rev. Soudure*, 1960, **16**, (1), 1–27) The phenomenon is described, methods of detection are discussed, the conditions in which micro-cracks appear, hydrogen in steel, metallographic observations, and theories of the phenomena with special reference to H_2 and to static fatigue are then considered.

Furnace for high-temperature brazing T. A.

Dickinson (*Ind. Gas.* 1959, **22,** Dec., 14, 16) A bell-jar type furnace heated by natural gas is described.

Brazing by electric resistance furnace. Principles and factors of brazing J. Guingand (*Usine Nouv.*, 1960, **7,** Feb. 18, 101–102; *Bulletin of Information Heurtey*, No.14) A technical article, with diagrams showing (1) ten examples of brazing and (2) eight contrasted examples of incorrect and correct methods. Equipment, the use of four classes of gases, particularly the use of combustible gas in the brazing of mild steel to copper, and methods generally to be followed are described.—S.H.-S.

Metal adhesives. and their application in bonded sandwich structural forms H. R. Ashley (*Manch. Assoc. Eng.*, 1959–60, (4), 1–12) Mainly non-ferrous.

Adhesives up diecasting output (*Iron Age*, 1959, **184,** Nov. 26, 94–95) An epoxy resin adhesive is being used to bond three separate die castings into one complex pump assembly. No lugs or flanges are required and close tolerance machining of mating surfaces is said to be unnecessary. The bonding operation can be performed by unskilled labour and results in an over-lap shear strength of 4670 psi.

General recommendations for the oxygen cutting of mild or low-alloy steels (*Cien. Tecn. Solda.*, 1959, **9,** Nov.–Dec., 9, 942, 1).

Flame cutting material 5 ft thick British Oxygen Gases Ltd (*Engineering*, 1959, **188,** Dec., 4, 595) An account of a demonstration given at Osnabruk.

O_2 cutting keeps pace with progress H. Thomasson (*Canad. Metalw.*, 1959, **22,** April, 54–56) Recent developments in O_2 cutting are surveyed. These include: more multiple torches; more mechanical or automatic handling of plates and parts; pattern-tracers using 'electric eyes' or model-size templates; application to high carbon and alloy steel plates.

Plasma—a substitute for the oxy-fuel flame J. A. Browning (*Weld. J.*, 1959, **38,** Sept., 870–875) The theory and applications of plasma flames are discussed, and an important future is forecast due to the very high temp. attainable in this way. Welding, metal-cutting, and spraying of refractories and substances that are difficult or impossible to spray by conventional means are considered. and the savings in cost and time by the use of plasma flames are pointed out.

Gas, arc cutting simplified J. Okladek (*Canad. Metalw.*, 1959, **22,** Dec., 18–23) The practices of gas cutting, lancing, scarfing, and gouging are considered and extended to heavy cutting, hot cutting, and under-water cutting. A useful table called 'metal cutting operating conditions' is appended.—T.G.

MACHINING AND MACHINABILITY

Machinability O. Andersson (*Tek. Tidsk.*, 1960, **90,** March, 18, 273–282) Methods of assessing machinability are reviewed, methods of classification discussed, and the effects of composition, structure, and heat treatment on the machinability of various materials is considered.

A three-dimensional, tool-life equation—machining economics B. N. Colding (*Trans. ASME, J. Ind.*, 1959, **81B,** Aug., 239–250) Two tool-life equations, one limited and one general, are derived, between the variables cutting speed, chip equivalent (introduced by Woxén), and tool life. A relationship is then derived which is used with the general tool-life equation to investigate the optimum combination of tool life, chip equivalent, and cutting speed (21 refs).

Some observations on the shearing process in metal cutting S. Kobayashi and E. G. Thomsen (*Trans. ASME, J. Eng. Ind.*, 1959, **81B,** Aug., 251–262) It was established that the shearing stresses on the shear plane for SAE 1112 free-cutting steel appeared to be independent of normal stress, finite strain, strain rate, and initial state of cold work under the conditions investigated (29 refs).

Observations on the angle relationships in metal cutting D. M. Eggleston, R. Herzog, and E. G. Thomsen (*Trans. ASME, J. Eng. Ind.*, 1959, **81B,** Aug., 263–279) Orthogonal-cutting experiments with SAE 1112 free-cutting steel using 18–4–1 high speed steel tools showed under the conditions of the experiments that the angle λ between shear plane and resultant tool force R was only approximately constant for each test condition and varied with cutting speed. The implication of these findings are discussed (16 refs).

Highlights on the Ford Anglia line-3 W. A. Hawkins and K. Hards (*Metalw. Prodn.*, 1960, **104,** March 16, 463–466) An illustrated account of the installation of broaching machines for machining bearing caps in a nine-station line.

Ultra-high tensile steels are machined hot R. A. Scott (*Metalw. Prodn.*, 1960, **104,** March 16, 484–485) Preliminary work with carbide tools shows the advantages of machining hot. Work up to 1200°F is being evaluated.

Mill quality molds with tracer Pacific Mold Inc. (*Iron Age*, 1959, **184,** Nov. 26, 101) An account is given of the production of complex moulds by milling controlled by a fully automatic three-dimensional tracing attachment following the contours of a plastic or soft wood master.—A.G.

Transfer line speeds structural fabricating Mulfab. Corp. (*Steel*, 1959, **145,** Jan. 11, **146.** 70–71) The mechanized facilities in a shop engaged in cutting and drilling beams are described. A craneman, a sawing line operator and a drilling line operator now do more work than was previously done by seven men.

Facing units make special machines Clay Cross (Iron & Foundries) Ltd (*Metalw. Prodn.*, 1959, **103,** Dec. 30, 2095–2097) Standard units are used to make up a flange facing machine for an iron pipe machining line.

Some metallurgy for the grinding experts T. Curti-Jung (*Met.-Reinigung Vorebhand.*, 1960, **9,** Feb., 26–27) Some notes on definitions and properties of irons and steels.

Precision barrel deburring of camshaft gear Vauxhall Motors Ltd (*Metalw. Prodn.*, 1960, **104,** March 16, 477–479) An illustrated account of a plant with an output of 64 shafts/h.

Ultra-high strength alloys Part I: Machining tests using high-speed steels and tungsten-carbide tooling J. Maranchik jun., J. V. Gould, and P. R. Arzt (*Aircraft Prod.*, 1959, **21,** Nov., 402–409) The relation of tool-geometry to speed and feed is included.

High output machine tools for cutting of metals K. W. Michler (*Metall.*, 1960, **14,** Apr., 318–322) The latest trend in machine tool development is a deviation from the transfer machine towards a high-output machine using cemented carbide or ceramic tool tips to give the high cutting speed required. The author describes a number of universal machine tools which can be adapted to automatic or partly automatic operation. The advantages offered by these are enumerated.—T.G.

Heavy cuts with carbide at low speeds (*Metalw. Prodn.*, 1960, **104,** March 9, 426–429) An illustrated account of the uses of Ardoloy grades.

Ceramic cutting tools P. T. Fitzroy (*Commonw. Eng.*, 1960, **47,** Jan. 5, 24–27) Work carried out at Melbourne is reported; tool properties and operating conditions are given and the effect of cutting speed on wear pattern and tool life.

Ceramic tools: Let them work where other materials fail H. Frommelt (*Iron Age*, 1959, **184,** Oct. 15, 106–108) The field of application of ceramic cutting tools is discussed with some examples. These tools are said to turn 63 Rc material up to 350 sfpm and to produce a superior finish.—A.G.

Now—successful ceramic milling R. M. Gill and C. Spence (*Metalw. Prodn.*, 1960, **104,** March 23, 514–517) The use of ceramic tipped milling cutters by Alvis Ltd is described.

Drill points modified for tough steels J. A. Edwards (*Metalw. Prodn.*, 1959, **103,** Dec. 30, 2112) A modified notched point is described.

Characteristics of spark erosion circuits P. G. Farley (*Met. Treatment*, 1959, **26,** Nov., 405–420) The theoretical basis of spark-machining circuits is considered. Circuits based on dc and a capacitance and on ac are described as well as pulse-forming circuits.—T.G.

Spark machining and its development D. W. Rudorff (*Met. Treatment*, 1960, **27,** Jan., 3–14) The nature of the discharge, the surface finish obtained, development of the circuit, electrodes, feed control, and future developments are considered.

The electro-spark machining of wire-drawing dies H. G. German (*Wire Prod.*, 1959, **7,** March–April, 1–8) The theory and practice of the electro-spark machining of wire drawing-dies are explained. In particular the 'G.K.N.' spark machine is described; this has been designed to allow practical economic operation. Dies can be made in hardened materials with a good accuracy and surface finish. Seven illustrated examples are given.—D.L.C.P.

Latest progress in the field of electro-erosion machining Y. S. Kroundycheff (*Swiss Techn.*, 1959, **2,** 49–51) After mention of two principal methods of electro-erosion, a third process, claimed as giving great stability and versatility, in use at the Charmilles Engineering Works Ltd of Geneva by their Elerodа D1 machine is illustrated and described.—S.H.-S.

Study of cutting fluids, III. On the roughness of machined surfaces N. Shinozaki and M. Harada (*Rep. Inst. Phys. Chem. Res. Tokyo*, 1959, **35,** March, 140–148) [In Japanese].

Automatic chip handling, oil recovery pay off Link-Belt Co. (*Steel*, 1960, **146,** Jan. 18, 90–91) The working and advantages of a metal-chip handling system which produces oil-free chips of uniform size is described. The turnings are crushed so that they have a higher bulk density and are easily handled. Oil extraction is effected in a centrifuge; the re-claimed oil is purified and re-used.—D.L.C.P.

Automation of chip removal at G.E. (*Canad. Metalw.*, 1959, **22,** May, 26–28) The chip disposal system installed for heavy duty lathes is described. The conveyors are of the hinged-pan type, which work without wedging or spilling.—D.L.C.P.

Electrolytic machining advances (*Steel*, 1959, **145,** Nov. 9, 156–157) A new 'Anocut' electrolytic cavity sinking machine and its applications are described. A shaped electrode is made negative and the workpiece positive in a harmless electrolyte solution. Material is removed at up to 0·005 in. per sec. to form open or blind holes. Good tolerances and finish can be maintained. The principal application is working high strength steels for special purposes such as jet engine parts.—D.L.C.P.

Cutting and metal-working oils W. R. Steet, F. Rogerson, and L. A. Cooper (*J. Jun. Inst. Eng.*, 1959, **70,** Oct., 13–23) A general review of the lubrication of workpieces of various metals for various machining operations.

The pretreatment is important; K. A. van Oeteren-Panhäuser (*Met.-Rein, Verbeh.*, 1959, **8,** Aug., 129–132) Several examples from practice are discussed to show the importance of suitable pre-treatment in preventing defects in the finished articles.

CLEANING AND PICKLING

New experiences in metal cleaning A. Pollack (*Met.-Rein. Vorbeh.*, 1960, **9,** Feb., 21–25) Topics dealt with in outline include: vapour and vapour jet degreasing, ultrasonic cleaning, economic advantages of alkali degreasing, new equipment for electrolytic cleaning and the future prospects of automatic cleaning plants. Some illustrations of equipment are reproduced.

20,000 Degreasing plants have been made by I.C.I. (*Ind. Fin.*, 1959, **11,** Aug., 54–56) Trichlorethylene plants are referred to.

Pretreatments required by protective coatings B. Waester (*Ind. Fin.*, 1959, **11,** Dec., 36–38; from *Kunststoffe Plastics*, 1959, 532–536) The necessity for a cleaning pre-treatment of metal parts before the application of protective coatings is discussed, and the several methods used are described. The results of tests made to determine the relative effects of these methods are given. Although the data are insufficient for certain evaluation, CS_2 degreasing appears especially favourable in the case of carbon steel.—G.F.

Speedy, precise, economic ultrasonic cleaning of metal parts (*Ind. Fin.*, 1959, **11,** Dec., 34–36) The principle of ultrasonic cleaning is outlined, and the design and operation of typical equipment are described. The advantages of this method over other cleaning processes are

discussed, and the direct and indirect savings resulting from its use are listed.—G.F.

A review of: barrelling and abrasive blasting (*Corros. Prev.*, 1959, **6**, 61–67) A review of processes, equipment, and application, including Roto-Finish and Vaquablast plant and blasting, liquid honing, and descaling systems.

Barrel finishing: materials and techniques for precision results D. Mosher, F. Soper, and T. Loll (*Products Fin.*, 1960, **24**, Jan., 27–31) Media, abrasives, speed, cycle time, and barrel size are discussed.

Vibratory motion gives faster barrelling (*Ind. Fin.*, 1959, **11**, Aug., 42, 60) The performance is outlined of a range of metal finishing machines developed by Roto-Finish Co. of Kalamazoo, USA, in which vibratory motion is used instead of the rotary action of the conventional barrel.

Barrelling and drum polishing G. Rousselot (*Galvano*, 1959, **26**, Nov., 513–514) A brief review of equipment, its use for cleaning of cast, forged, and stamped parts, and for surface improvement, the choice of abrasive additive, rate of rotation, and time of treatment and size of charge is given.

Metal abrasive producers put accent on performance (*Steel*, 1960, **146**, Feb. 1, 88–90) The views of makers of metal abrasives for blast cleaning and peening, on products and met hods are reviewed.—D.L.C.P.

Abrasive compounds in barrel finishing A. S. Kohler (*Plating*, 1960, **47**, Jan., 59–62) Deburring compounds with and without abrasive grains are reviewed, considering particle size and wear and their advantages and disadvantages.

Problems of dressing with metal shot-blasting agents H. Thier (*Met.-Rein., Vorbeh.*, 1959, **8**, Sept., 145–150) Some experiences are described of the use of various shot-blasting materials in the preparation of mild-steel sheet work for enamelling.

Blast cleaning machine speeds scale removal Vulcan Rivet & Bolt Corp. (*Steel*, 1959, **145**, Dec. 7, 159) The installation of a blast cleaning machine for nut and bolt blanks gave the following advantages over the previous system: faster, cleaner, reduced maintenance, cleaner and less damaged product.—D.L.C.P.

Post-treatment solves problems in work with shot-blast stock B. Pais and E. S. Keely (*Iron Age*, 1959, **184**, Nov. 5, 94–96) Means for overcoming some problems in the shot-blasting of sheet and plate steel are described. These problems include residual fine scale (smut), dust extraction, and die wear due to the increased hardness of the sheet.—A.G.

Testing and standardization of steel shot E. Bickel and H. Krautmacher (*Stahl Eisen*, 1960, **80**, March 17, 365–367) Two letters. One discusses a paper recently published and points out that the shot-peening effect and the cleaning effect in shot blasting are two entirely different things that should not be confused. The previous author replies and stresses *inter alia* that the use of small-size shot has three advantages: (i) the wear of the shot-blasting machine is reduced; (ii) the service life of the shot is greatly improved; and (iii) the abrasion of the parent metal below the scale is kept smaller.—T.G.

Abrasive blasting in the shipping industry P. B. Wharton (*Corros. Prev.*, 1959, **6**, June, 59–60) Modern developments in sand blasting for marine applications are described.

You can cut blast cleaning costs G. O. Pfaff (*Foundry*, 1959, **87**, Oct., 92–95) Measures taken to check blast cleaning efficiency and costing are described.

Fettling costs reduced: change-over to electric grinders at Bradford foundry English Electric Co. Ltd (*Iron Steel*, 1960, **33**, Jan., 21–22) Operation of and economies obtained by the use of high-frequency electric grinders at the Dick Lane foundry are described.

Some studies on the shot descaling of rolled wire and hot strip H. Sedlaczek, H. Wirtz, and M. Buch (*Stahl Eisen*, 1960, **80**, March 17, 351–356) Experiments on shot descaling of rolled steel wire showed that the shot blasting machines must have variable speed control in order to deal with wire of different surface conditions. It was found also that the quantity of shot used per unit of time and the size of the shot have a marked effect on the surface quality of the treated wire. In the studies on the shot descaling of hot-rolled strip it was found that clogging up of the rolls in the subsequent cold rolling could be drastically reduced if the edges of the strip were properly descaled.—T.G.

Surface finishing with felt grinding and polishing materials T. Curti–Jung (*Met.-Tein. Vorbeh.*, 9160, **9**, March, 39–41) The materials and the finish produced are reviewed.

Automatic polishing with liquid composition Vauxhall Motors Ltd (*Electropl. Met. Fin.*, 1960, **13**, April, 130–132, 138) An illustrated account describing the use of the Acme Universal and the Morrisflex–Hammond machines.

Flame-priming improves metal painting G. H. Morgan (*Eng. Min. J.*, 1959, **160**, April, 86, 88) Spaced oxy-acetylene flames are used, removing mill scale and converting rust to stable oxides by loss of water.

Flame-cleaning: surface preparation systems combined on 'live' gas holder West Midlands Gas Board (*Corros. Techn.*, 1960, **7**, Feb., 54, 64) A description of the flame cleaning of a gasholder containing gas, using a strip cleaning technique combined with ancillary descaling equipment.

Gas-boosted air jet speeds blast cleaning J. A. Browning (*Metalw. Prodn.*, 1960, **104**, Feb. 17, 293–294) Blast from burning gas and compressed air provides the high velocity of the air jet in a novel sandblasting process that is claimed to be four times as effective as conventional methods.—T.G.

Blast descaling reduces pickling cost Universal Cyclops Steel Corp. (*Steel*, 1959, **135**, Nov. 16, 118–122) Details of a plant are given in which a shotblasting machine removes scale from a stainless steel strip before it is pickled This pickling time and process costs are reduced.—D.L.C.P.

Electropolishing of stainless steels (*Techn. Indust.*, 1959, **37**, Nov., 1662–1664) [In Spanish] The techniques and characteristics of electro-polishing are described; examples of its applications in industry are given.—P.S.

Electropolishing techniques for metals and alloys S. Gruettner (*Ind. Fin.*, 1959, **11**, Aug., 49–50, 52; from *Metallwarenindustrie und Galvanotechnik*, **49**, (2), 58) The history of the development of electropolishing is given and the article describes the techniques used.

Modern continuous pickling processes H. Moehring (*Met.-Rein. Vorbeh.*, 1960, **9**, Jan., 3–9) A general description of modern continuous processes of pickling, degreasing, and washing is given, with particular reference to treatments prior to enamelling.

Some investigations into the acid pickling of mild steel D. Jackson, A. J. Stedman, R. V. Riley (*Proc. Inst. Virt. Enamellers*, 1954–1957, **12**, 79–92) Experimental work carried out to compare the characteristics of 10%H_2SO_4 and 20%HCl as mild steel pickling agents and to examine the effect of several variables is first described, followed by an account of research on the effects of previous alkaline cleaning in the pickling process. The paper concludes with data obtained in a study of the Ni-dip process and on work carried out to improve the surface condition of steel after pickling.

Electrolytic pickling of sheets by mains frequency alternating current P. P. Belyaev, B. A. Nikitin, S. I. Sukhov and (*Stal'*, 1960, (1), 79–81) Graphite electrodes in HCl solution were used. Cl_2 is not evolved. The method was successful for bright-annealed hot-rolled and cold-rolled sheets. As no gases are evolved the installation can be built into the hot tinplating line.

Sensitivity of different steels to pickling brittleness C. A. Zapffe and M. E. Haslem (*AD*-137777; *PB* 136226, 1948, pp.27; from *US Res. Rep.*, 1959, **31**, 249) In cathodic pickling annealed steel is least and hardened steel most susceptible and cold working increases susceptibility. Martensitic stainless grades are most susceptible and this is increased by increased C. In acid pickling stainless steels are less susceptible except for 440–C.

Inhibition of hydrogen embrittlement during pickling R. C. Bertelson (*PB* 140306, 1959, Feb., pp.13; from *US Res. Rep.*, 1959, **32**, Aug. 14, 238) A synthesis of 6-methyl-3-p-tolyl-3: 4-dihydroquinazoline which is a component of US Army 4-105 inhibitor giving 53% yield is worked out.

Effluent testing station (*Wire*, 1959, June, 64–66) Design features and operating data of an automatic effluent sampler for a testing station are described.

The problem of industrial waste waters (*Usine Nouv.*, 1960, **16**, Feb. 25, 26–27) A review of two papers presented before the Institute of Sanitary Techniques on Dec. 14 1959. (1) R. Coles states a case for **The role and formation of a body of water specialists, 'water-engineers'. as yet non-existent**, (2) M. A. Celerier, **The problem of industrial discharges: and attempt at synthesis**, in which methods of treatment are suggested and briefly described.

Treatment of waste acid and effluent from continuous strip pickling lines A. E. Matthews (*Wire Ind.*, 1959, **26**, May, 453–459, 492) The new plant for treating waste acid and effluent from the continuous strip pickling lines at the Brinsworth Mill is described. The plant was designed to run with the minimum attention. Regeneration of the spent acid is by the shock cooling process; 'Dynacone' continuous centrifuges separate the ferrous sulphate crystals from the regenerated acid. Rinse water from the pickling lines is neutralized by the addition of lime slurry.—D.L.C.P.

PROTECTIVE COATINGS

Some recent developments in protective coatings in U.K. A. N. Kapoor (*NML Tech. J.*, 1959, **1**, Nov., 16–27) The author describes aluminizing practice at Bridgend and tinning practice at Velindre.—T.G.

Russian practice in electroplating steel N. P. Fedot'ev and S. Ya. Grilikhes (*Electropl. Met. Fin.*, 1960, **13**, April, 119–125) This is translated from a Russian book 'Electropolishing, Anodizing and Electrolytic pickling of metals'. The extract given covers electropolishing bath compositions and operation, electrolyte regeneration, and a list of defects and their remedies.

Metal finishing in Europe today D. A. Marino (*Met. Fin.*, 1959, **57**, Dec., 54–58) A general account, with particular reference to France and Italy.

Problems and trends in modern electroplating (*Galvano*, 1960, **29**, Jan., 37–40) The final article of a series (45 refs).

Specifications for protective plating: 2 E. A. Ollard (*Corr. Prev.*, 1959, **6**, Sept., 42–47) The uses are considered, mainly for small objects and mostly non-ferrous.

The preparation of electrolytic baths H. Richter (*Galvano*, 1960, **29**, Jan., 19–23) The quality of the materials including the water and the making up of cyanide baths and alkaline Zn and brass plating baths are considered.

Testing of protective coatings for metals W. E. Kirkendall (*Corros. Techn.*, 1960, Feb., 41–45; March, 88–91) This first article deals with the principles and application of newly-developed indicator solution and electrolytic conductance tests on protective coatings for steel pipe, with the primary object of supplying comparative data on performance characteristics of tape-wrap coatings compared with coal-tar enamels and other coatings. In the second part, practical methods are described for obtaining laboratory data on the performance of protective coatings, including plastic tape-wrap coatings applied to steel pipe.

Testing plating durability: the copper-accelerated acetic acid test G. L. Sukes (*Met. Fin.*, 1959, **57**, Dec., 59–65) The development and application of the Cass test are described, and its reliability discussed.

Tests of electrodeposits (*Galvano*, 1959, **28**, Sept., 418–422) Chemical, metallographic, and magnetic examination for thickness measurement are reviewed with a brief survey of US measuring devices. Etching methods are given in some detail.

The testing and examination of electrodeposits, II R. Quarendon (*Product Fin.*, 1960, **13**, Jan., 85–88) The measurement of lustre and brightness of electrodeposits is reviewed and a number of instruments for this purpose is

described. **Part 3** (Feb., 91–96) Glossmeter techniques are reviewed.

Rough surface aids bonding between dissimilar metals National Lead Co. (*Iron Age*, 1959, **184,** Nov. 5, 90–91) In the bi-metallic interlock bonding method described, cast iron is mechanically bonded to Al using a die-casting process. By the use of a rough cast-iron surface a bond is obtained which is said to resist high temp. and shear stresses.—A.G.

A new economic method of treatment of stainless steels prior to cadmium plating from conventional cyanide solutions L. Domnikov (*Plating*, 1959, **46,** May, 481–483) A new electroless method of preparing stainless steel surfaces for cyanide-bath Cd plating is described. This involves (1) an alkali-acid cleaning (2) treatment in 35% H_2SO_4 (3) Ni flash in immersion bath (4) cyanide dip.—D.L.C.P.

Aluminium bonded by diecasting process (*Steel*, 1959, **145,** Nov. 30, 98–100) A technique for bonding Al to cast iron using a diecasting process is described. The iron casting only requires simple cleaning, and the Al is forced into the rough surface giving a 'mechanical interlock'. A permanent stable joint is obtained. Examples of applications are: cast iron cylinder liners for Al engine blocks; cast iron brake liners for Al drums and wheels; Al cooling fins on cast iron piston sleeves.—D.L.C.P.

Thin metal strips for sheathing telephone communication cables H. N. Padowicz (*Wire Wire Prod.*, 1959, **34,** Nov., 1482–1485, 1548–1550) The thin metal strip components of the 'Alpeth' and 'Stalpeth' telephone cables manufactured by Western Electric Co. are described. These cables dispense with load sheathing and incorporate Al and Al/steel strips respectively. The technique of soldering the steel strip is explained.—D.L.C.P.

Sherardizing: a method of protecting ferrous metals A. E. Williams (*Prod. Fin.*, 1960, **13,** March, 54–60) The process, mixtures used, advantages and limitations of the method are discussed.

Science for electroplaters. 53-Complex cyanide baths L. Serota (*Met. Fin.*, 1959, **57.** Dec., 75–78) Bath composition, deposition mechanism, and properties of cyanide baths for Cu plating are reviewed.

Metal spraying (*Met. Elect.*, 1959, **23,** Feb., 85–90; March, 96–99) [In Spanish] The means of spraying, the composition and structure of the sprayed metals, and the mode of contact between coating and base metal are discussed. Typical applications are described and include corrosion protection and renovation of worn parts.—P.S.

Metal spraying under protective atmospheres W. E. Stanton (*Eng. Dig.*, 1959, **20,** Nov., 445–447) Spraying of iron and other metals in argon and the structure produced are described.

Contract metal spraying G. A. Curson (*Electropl. Met. Fin.*, 1960, **13,** March, 91–94) A review of gritblasting and spraying with Zn, Al, and other metals, hard facing, and the use of automatic plant.

The economies of vacuum metallizing M. B. Gilbert (*Prec. Met. Mold.*, 1960, **18,** Jan., 51–53) A general account of the process, equipment and tests.

The A–C resistance of a polarized stainless steel wire cathode G. M. Schmid and N. Hackerman (*J. Electrochem. Soc.*, 1960, **107,** Feb., 142–143) An attempt to measure the growth of the oxide film is described.

Let phosphates reduce friction Pneumatic Scale Corp. (*Iron Age*, 1959, **184,** Oct. 29, 127) A brief description is given of the Endurian process which coats machine parts with a complex phosphate. This surface is said to have excellent resistance to wear and corrosion an also to act as a lubricant.—A.G.

Clean, phosphatize and paint metal parts in compact unit E. I. Du Pont de Nemours & Co. (*Iron Age*, 1959, **184,** Nov. 5, 87–89; *Steel*, 1959, **145,** Nov. 2, 76–77) The unit described incorporates trichlorethylene cleaning and phosphatizing without the use of aqueous solutions and final painting. Advantages claimed include savings in cost and space, improved corrosion resistance, and the elimination of fire hazard during painting.—A.G.

Phosphating, electropolishing and high-speed chromium plating in buffer production H. L. Watts (*Electr. Met. Fin.*, 1960, **13,** Feb., 45–47) Cr plating of pistons is considered under the headings: hard-chromium plating conditions, plant, jigging, and cost savings.—T.G.

A new approach to vitreous enamelling (*Product Fin.*, 1960, **13,** Jan., 67–68) A low-pressure spray gun for spraying vitreous enamels is described. It is claimed that by spraying, considerable savings in the costs of finishing can be achieved.—T.G.

The past, present and future developments of vitreous enamel as a finish T. J. MacArthur (*Proc. Inst. Vitr. Enamellers*, 1954–1957, **12,** 119–122) The development of vitreous enamelling in recent years, is outlined, its position compared with other finishes, and possibilities of improving technology are reviewed.

Mechanization in enamelling: a symposium: pickling and pre-treatment J. Stack (*Proc. Inst. Vitr. Enamellers*, 1954–1957, **12,** 123–125, 129) An outline is given of mechanical developments used for chemical processes applied to sheet-steel articles prior to the application of vitreous enamel.

Mechanization in enamelling: a symposium: drying and fusing T. MacP. Lees (*Proc. Inst. Vitr. Enamellers*, 1954–1957, **12,** 130–138).

Some aspects of vitreous enamelling practice in Australia A. R. Williams (*Proc. Inst. Vitr. Enamellers*, 1954, 1957–**12,** 23–26, 78) Development of Australian practice, and typical control methods are reviewed.

Curtain walling with vitreous enamelled steel sheet T. J. MacArthur (*Ind. Fin.*, 1959, **11,** Aug., 36–38) A description of the materials and technique.

Vitreous enamel stops water corrosion in boilers J. F. Stevens (*Ind: Fin.*, 1959, **11,** Nov., 30–31) Stewart Gray Ltd, have developed a new vitreous enamel as a protective barrier on the cast iron of boiler waterways to resist the corrosive action of the water. The author describes the successful testing of the enamel in boiler manufacture.—G.F.

New enamels for application to sheet iron A. Nicholls (*Ind. Fin.*, 1959, **11,** Aug., 38) Recent improved titania self-opacified cover coat enamels, manufactured by Metal Porcelains Ltd of Smethwick, give very thin coatings with satisfactory properties.

The establishment of standards of quality for vitreous-enamelled ware J. M. Wijtenburg (*Proc. Inst. Vitr. Enamellers*, 1954–1957, **12,** 69–74) The practical usefulness of standards of quality in this field is discussed, and a method proposed for setting up quality standards and providing a suitable information system; a detailed example is given.

The development of opacity in white porcelain enamels A. I. Andrews (*Proc. Inst. Vitr. Enamellers*, 1954–1957, **12,** 60–64, 100) Factors influencing opacity are discussed, the development of opacifying additions is outlined, and typical composition factors in the development of modern white opaque enamels are tabulated.

An investigation into the effect of detergents on vitreous enamel Institute of Vitreous Enamellers Techn. Sub-Committee (*Vitr. Bull.*, 1960, **10,** Feb., 285–292, discussion 293–298) A simple testing method is proposed and nine proprietary preparations (named) and Na_2CO_3 were examined and the results are tabulated. Good correlation was obtained by different observers and degree of corrosion is indicated In all cases the attack was progressive with time of immersion.

Detergent resistance of enamels tested and improved F. G. Morriss (*Ind. Fin.*, 1959, **11,** Aug., 29–30) Types of test are compared, and the advantages of the ring method pointed out. Factors influencing resistance to detergent action are discussed; of these, increase in detergent temp. gives a marked increase in attack; other factors include mill additions, and firing time and temp.

Assessment of costs A. Biddulph (*Ind. Fin.*, 1959, **11,** Aug., 31) Some ways in which economies may be made in enamelling are enumerated.

Measurement of sub-sieve particle size in vitreous enamels J. F. Browne (*Met. Fin. J.*, 1960, **6,** Jan., 18–19, 27) The methods are listed and the Andreason pipette and hydrometer methods are regarded as practical and the latter is described.

An account of de-enamelling practice in a French works M. Camuset (*Proc. Inst. Vitr. Enamellers*, 1954, 1957–**12,** 75–78) A description of the de-enamelling plant at SODI, which uses fused NaOH.

Temporary coatings for corrosion protection D. H. Weatherley (*Corr. Prev.*, 1959, **6,** Sept., 36–40, 54) Strippable protective coatings and cocooning methods are reviewed.

Organic finishes for metal, VII B. M. Letsky (*Product Fin.*, 1960, **13,** Jan., 60–62) Roller-coating finishes on metal in the canning industry, a gold razor blade lacquer, and reverse roller-coating are dealt with.—T.G.

Organic coatings for sheet and strip metals E. E. V. Sharpe (*Sheet Met. Ind.*, 1960, **37,** Feb., 97–100) Continuous coating is outlined and paint priming and finishing and the use of PVC are briefly discussed.

Applications of plastic-coated steel sheet and strip W. N. Jenkins (*Sheet Met. Ind.*, 1960, **37,** Feb., 89–96) Availability, fabrication, and uses are reviewed.

Form decorated parts from coils J. J. Obrzut (*Iron Age*, 1959, **184,** Oct. 15, 101–103) The method of production of Miracoil, a pre-decorated steel strip protected by a transparent plastic coating, is described. Properties and applications are also discussed.—A.G.

Combine plastics and metals for better products (*Iron Age*, 1959, **184,** Nov. 5, 98–99) Some applications of plastic coated metals are discussed, including printed circuits, tin plate rolls, and ball-bearing retainers.—A.G.

Epoxy coating resists high gamma radiation: can be used on steel or concrete (*Mat. Design Eng.*, 1960, **51,** Feb., 9) An amine-cured epoxy coating is referred to, the resistance to γ-rays of test panels is mentioned.

The manufacture and applications of PTFE-lined pipes and components T. R. Thierry (*Corr. Prev. Control*, 1960, **7,** Jan., Supp. *Corros. Eng.*, ii–v) The properties, fabrication methods, and performance are briefly reviewed with notes on uses.

Organic coatings for steel drums D. J. S. Shaw, R. M. C. Logan, and H. B. Footner (*Ind. Fin.*, 1960, **12,** Feb., 29–33) Recent research in the packaging field has concentrated on achieving a working compromise between coating reliability and cost. Both external and internal coatings are considered, and the materials and methods used are discussed.

Protective coatings for drain pipes and water mains V. M. Pishchev (*Stal'*, 1960, (1), 62–66) An automatic production line for continuous coating is described. Petroleum bitumens are used, as the coal tar coatings so far employed are unsatisfactory in protective properties as well as toxic. The properties of tar and bitumen are compared and rates, temperatures, and air-cooling of the coatings are indicated.

Protection of steel and other materials with hard and soft rubber H. Ziegner (*Wire*, 1959, (43), Oct., 136–138) Applications of hard rubber to the lining of pipes and other components for the chemical and electrical industries are described, with some notes on techniques, and reference is made to the use of soft rubber where abrasion resistance is required. A table is given showing the resistance of hard and soft rubber to various chemicals.

Protective organic paint coatings J. H. Ousbey (*Corr. Prev. Control*, 1960, **7,** Jan., 26–32, 44–45) A review of synthetic resin and silicone vehicle paints.

The production and application of pre-painted sheet and strip I. R. Banks (*Sheet Met. Ind.*, 1960, **37,** Feb., 101–106) A typical strip-coating installation is outlined and methods of roller coating, baking, and recoiling are described. (**Discussion of preceding three papers.** 106–112).

Metallic lead as a pigment for anticorrosive paints J. A. Calcutt (*Corr. Prev.*, 1959, **6,** Nov., 34–38) With illustrations of large structures protected with Pb.

Rust inhibition through calcium plumbate pigment H. C. Wesson (*Corros. Techn.*, 1960, **7,** Feb., 37–38) Characteristics and applications of calcium plumbate as a paint primer are described.

Marine paints and compositions–I T. A. Banfield (*Corros. Prev.*, 1959, **6,** June, 51–55) The main differences between the painting of

ships' hulls and land structures are examined; this part deals with surface preparation, pretreatment, priming paints, and general considerations in relation to finishing paints.

Flow-coat finishing of steel roof deck. E. A Blount (*Products Fin.*, 1960, **24,** Jan., 20–26) Painting of the Granite City Steel Co. roof deck sections and guard rails is described.

Electric control of paint protection of metal L. Rivola (*Met. Ital.*, 1960, **52,** Jan., 15–18) The degree of protection afforded by paints can be measured using a square wave current. The electric resistivity in relation to time determined during corrosion gives a quantitative measurement of the life of the paint. The method can be applied in all the evaluation tests for paints on metals such as salt spray, high humidity, and alternate and continuous immersion.

CLAD SHEET AND HARD FACING

Nickel-clad steel and monel-clad steel (*Wiggin Nickel Alloys*, 1959, (51), 3–9) Properties of these clad sheets are given.

Experience with the deposit welding of steel rolls of blooming-slabbing rolling mills by the submerged arc process H. Stoll (*Stahl Eisen*, 1960, **80,** April 14, 483–491) The submerged-arc deposit welding of worn steel rolls for blooming-slabbing rolling mills is described, the merits of high-wear-resistant deposits are stressed and the economic position of the method is discussed. The deposit-welding process should be an economic proposition as long as the quality and particularly the service life of the present-day steel rolls are not materially improved.—T.G.

Restoration by facing of worn parts of a dredger S. Lundin (*ESAB Rev.*, 1959, (1), 13–15).

Improved steel resistance: Surface hardness increased (*Ind. Fin.*, 1960, **12,** Feb., 44–45) The chromizing process produces a stainless alloy surface layer as an integral part of a steel article. The Arkrom processes of carburizing are described and the characteristics of the surface layer are discussed. The hardness characteristics of some chromized steels are shown graphically.—G.F.

Chromize bolts and tubing with gaseous compound G. Becker (*Iron Age*, 1959, **184,** Nov. 19, 130–132) The process described for chromizing parts involves the use of a chromium halide gas. Factors affecting the suitability of the base material for chromizing are discussed together with the properties of the coating formed.—A.G.

Practical aspects of hard-chromium plating C. F. Corfe (*Electr. Met. Fin.*, 1960, **13,** Feb., 48–52, 69) Among the subjects discussed are hard-chromium plating of steel, including chromic and sulphuric acid etch, and moulding dies. The removal of magnetic particles from Cr plating solutions is mentioned. The major part of the paper deals with plating of nonferrous metals.—T.G.

[**Chromium sheet production**] R. L. Samuel (*Corr. Prev.*, 1959, **6.** Nov., 44, 46) A letter. Questions about the Diocrom process are raised.

Refractory metal facings for protection of metal surfaces subjected to repeated high temperature pulses A. Cohen and E. Homer (*Trans. ASME(D)*, 1959, **81,** June, 197–202) The development of a facing to withstand the high temp. which may be developed in the resonant structure of a magnetron is described. An analytical determination of the temp. was made and experiments were conducted to verify the conclusions. Consequently, W-faced vanes were tried, and proved successful. Cu-impregnated W might prove an even more durable material.—D.L.C.P.

Heat-plating: A new process for the deposition of tungsten carbide and nickel and for hardfacing (*Met. Din. J.*, 1960, **6,** Feb., 49–50, 54) An account of the Pyro-tung (for WC) pyro-Plate for (Ni) and Nikl-Kase (hardening process) methods new used by Metallurgical Consultants Inc.

Flame spraying for ceramic coatings (*Ceramics*, 1959, **10,** Nov., 12–13) An account of the ThermoSpray process.

Ceramic coatings for use at high temperatures B. Zick (*Proc. Inst. Vitreous Enamellers*, 1954–1957, **12,** 65–68, 78) A review of work carried out, particularly by US National Bureau of Standards, on the development of ceramic coatings on steel (a) as heat-resistant coatings on mild steels (b) as thin-layer coatings to provide radiation suppression or reflection, or thermal insulation in specific high-temp. installations.

POWDER METALLURGY

Review of selected papers relating to new structural materials developed by powder metallurgy (*AD–75224*, 1955, Feb. 25, pp.29; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1199) A review of selected papers.

The forming of complicated shapes by powder metallurgy slip casting H. H. Hausner (*NYO–8665*, 1958, Aug., pp.20; from *US Res. Rep.*, 1959, **31,** March 13, 179; no abstract).

Symposium on the preparation and characteristics of metal powders Sylvania Electric Products Inc. (*TID–5153*, 1957, May, pp.194; from *US Res. Rep.*, 1959, **31,** April 17, 274; No abstract).

Powder metallurgy, Stainless steel-boron carbide and zirconium-boron-carbide compacts, Progress report P. G. Herold (*KAPL–M–JMC–1*, 1955, May, pp.29; from *US Res. Rep.*, 1959, **31,** April 17, 273; No abstract).

Modified H-iron process produces metal powder Alan Wood Steel Co. (*Steel*, 1959, **145,** Dec. 21, 94–96) Plant for producing iron powder from scale or high grade ore is described. H_2 is produced by reaction between coke oven gas and O_2; raw materials are prepared by purification and grinding; reduction is carried out in three fluidized beds at up to 500 psi and 1000°F. The product is then heated to high temp. in a non-oxidizing or reducing atmosphere and may be used for powder metallurgy or as melting stock.—D.L.C.P.

Recovery of iron powder by distillation of iron amalgam G. Jangg, G. Csizi, and H. Hohn (*Z. Metallk.*, 1959, **50,** Aug., 460–465) The method used to obtain a high quality iron powder from Fe amalgam, and the properties and testing of the powder are described. The process can be made continuous, and the material is more suitable for sintering than electrolytic Fe powder.

Steel bonded carbide materials Chromalloy Corp. (*Machinery*, 1960, **96,** March 9, 531) The Ferro-Tic S–45 and S–55 materials are described which are machinable, resistant to heat and corrosion and have the hardness of TiC.

Powder metallurgy develops a new cutting tool composition M. Ference, jun. (*Prec. Met. Mold.*, 1960, **18,** Jan., 42) A composition with 80%TiC, 10%Ni, and 10%Mo has been developed and tool performance data are indicated.

Iron–carbon system in powder metallurgy, Part I, Quality of iron powder and graphite P. U. Gummeson (*Prec. Met. Mold.*, 1959, **17,** Nov., 56, 58, 59) Graphite quality is considered in particular. A table of the properties of sintered iron–graphite compositions is given.

Measurement of the average equivalent diameter of certain powders by means of the air permeability of the layer U. Ts. Andres (*Zavods. Lab.*, 1959, (11), 1399–1400) The article examines P. C. Carman's (*ASTM symposium of new methods for practical size determination in the subsieve range*, NY, 1941) deductions and gives the theoretical basis for the determination. On the strength of experiments carried out, the author recommends this method, for rapid determinations of the average equivalent diameter of magnetite, ferrosilicon, and baryta powders.

Contours of metallic powders Ya. E. Geguzin and N. N. Ovcharenko (*Fiz. Met. Metalloved.*, 1959, (1), 38–44) The surface of the compacted copper powder objects annealed in an atmosphere of copper vapour, assumes an intricate contour consisting of macroscopic projections and depressions of about 100 μ diameter, each of which is endowed with a rough surface. The main cause of the formation of these contours is the vaporization of the metal with the subsequent condensation of the vapours on the surface of the specimen.

Some mechanical requirements of plant for the roll-compacting process A. F. Marshall (*Powder Met.*, 1960, (5), 24–31) A literature survey with discussion of type of mill, roll loads, lubrication, and other factors.

Continuous compaction by cyclic pressing C. Deibel, D. R. Thornburg, and F. Emley (*Powder Met.*, 1960, (5), 32–44) A process for producing bars, wide and thick and of unlimited length is investigated. Pressing, sintering, and subsequent working are studied.

Experiments with direct observation of the sintering process by high-temperature metallography E. B. Simonsen (*Planseeber.*, 1960, **8,** April, 2–13) The changes taking place during sintering of coarse-grain copper powder were observed in a high-temp. microscope. Temp. of up to 885°C were used. Subsequent microexamination showed that sintering started at a large number of contact points, thus forming pores or cavities which impede the progress of sintering. The experiments, including the hot-stage microscope, are described in detail.

Present position of the powder metallurgical finishing of sintered iron mouldings and sintered metal bearings T. Brüggemann and G. Unger (*Neue Hütte*, 1960, **5,** Feb., 118–121) Processes used and types of product in the powder metallurgical section of VEB Eisen- und Hüttenwerke Thale are outlined.

Automatic control of endothermic sintering atmospheres G. F. Sommer (*Metal Powder Ind. Fed.*, 15th *Annual Meeting*, 1959, 41–48) Furnaces for compact sintering are considered.

Fundamentals of sintering, III: Third Annual progress report for June 30, 1957 to June 30, 1958 A. L. Pranatis (*SEP–250*, 1958, June, pp.28; from *US Res. Rep.*, 1959, **31,** March 13, 179; No abstract).

The continuous sintering of copper—lead to steel P. G. Forrester and J. K. Beddow (*Powder Met.*, 1960, (5), 149–154) Bimetal strip is produced by sintering uncompacted powder to the steel, compacting by rolling, and resintering. The mechanism of the process is discussed.

Sintered stainless steel, I: the influence of alloy composition upon compacting and sintering behaviour, II: the properties of stainless steel powders sintered in dissociated ammonia R. L. Sands and J. F. Watkinson (*Powder Met.*, 1960, (5), 85–104; 105–115) Six stainless steel powders of different compositions were compacted and sintered under various conditions. Compositions selected from the first series of trials were treated under conditions simulating industrial practice and the properties of the products are given.

Effect of powder particle size on the grain size of the sintered material H. H. Hausner and R. King (*Planseeber.*, 1960, **8,** April, 28–36) Grain growth in the sintering of metal powder compacts is determined by the temp. and time of sintering. The rate of grain growth is, however, delayed by the pores between the bridges formed by the sintered grains. The authors studied the changes which the pores undergo during sintering.—T.G.

Processes in the compacting of metal powders to complicated parts H. Silbereisen (*Planseeber*, 1959, **7,** Aug., 67–78) It is pointed out that purely theoretical considerations are not of much help in discussing these processes, which must be followed by practical trials. Causes of cracking and exfoliation are considered, and the various stages in compacting simple and complicated parts are studied by examples.

Infiltration of iron powder with tin–lead alloy T. Kimura (*Planseeber*, 1959, **7,** Aug., 50–66) The mechanical properties were examined of synthetic alloys produced by the infiltration of spherical close-packed powders with a soft matrix, carbonyl iron, and a 50%Pb 50%Sn solder being used. The infiltration process was followed quantitatively with a dilatometer. Gases adsorbed on the Fe powder surface were found to be the main factor inhibiting infiltration. The mechanism of infiltration is discussed.

Nickel–iron magnetic strip by metal powder rolling W. M. Hubbard, J. F. Haben, and E. Adams (*NAVORD Rept. No.6257; PB* 140384, 1959, Jan., pp.25; from *US Res. Rep.*, 1959, **32,** Aug. 14, 238).

Study of sintering of carbonyl iron by electrochemical potential B. Bovarnick (*Planseeber.*, 1959, **7,** Aug., 34–50) The electrochemical

potential was measured of carbonyl iron pressed from 5·0 g/ml to 6·5 g/ml and sintered from 800 to 1200°C for various periods from ½–8 h. Values obtained were 720 mV for all unsintered specimens and 750 mV for lightly to 620 mV for severely sintered samples, corresponding tensile strengths varying from 10000–30000 lb/in². An expression was derived relating electrochemical potential to tensile strength (22 refs).

How fibres are spun from metal melts (*Steel*, 1960, **147**, Jan. 11, 74–76) Information from patent literature on recent developments in methods of making metal 'fibres' is presented. These methods employ synthetic textile techniques, i.e. extrusion or casting from liquid. Molten metal is forced through apertures and the stream cooled in various ways, e.g. on a rotating cup, on a continuous belt, or in a gas stream; the conditions can be adjusted to give various shapes, sizes, and forms of product. Material may be from a wide range of ferrous or non-ferrous alloys, or may be bi-metallic. Potential applications of the products are in filters, diffusers, mixers, flame barriers, battery plates, or structural uses.—D.L.C.P.

FERRITES, CERMETS, AND CARBIDES

Chromium carbide as a basis for hard metals having special properties V. V. Grigor'eva, V. N. Klimenko, and Ya. T. Kosolapova (*Voprosy Porosk, Metallurgii i Prochnosti Materialov: published by Ukrainian Academy of Sciences, Kiev*, 1958; 79–89) Properties of known Cr carbides are enumerated and compared with those of other metal carbides, and the production, properties, and possible applications of Cr_3C_2 are described (11 refs).

On the specific heat of MnZn- and NiZn-ferrite between 20°C and 350°C J. L. Verhaeghe, G. G. Robbrecht, and W. M. Bruynooghe (*Appl. Sci. Res.*, 1960, **8B**, (2), 128–134).

Magnetic properties of nickel-iron ferrite N. Menyuk and K. Dwight (*J. Appl. Phys.*, 1959, **30**, April, 36s–37s).

Phase diagram of the system BaO–Fe_2O_3 Y. Goto and T. Takada (*J. Amer. Ceram. Soc.*, 1960, **43**, March, 150–153).

The thermomagnetic behaviour of pure nickel ferrite L. F. Bates and H. Clow (*Proc. Phys. Soc.*, 1960, **75**, Jan., 17–23).

The influence of Fe_2O_3-materials on the production of Mn–Mg ferrites A. Bergstein, M. Rozsival, and M. Mikulaš (*Silikáty*, 1960, **4**, (1), 60–65).

Diffusion lag in nickel–zinc ferrites rich in iron A. Marais and T. Merceron (*Compt. Rend.*, 1959, **249**, Dec. 9, 2511–2513) Permeability and loss factor show three bands between liquid N_2 temp. and the Curie point. Mechanisms are indicated.

Remagnetization experiments in $Mn_{1+x}Fe_{2-x}O_4$ T. J. Matcovich and C. J. Kriessman (*J. Appl. Phys.*, 1959, **30**, April, 26s–27s).

Resolution of the ambiguity of valence states in spinels containing manganese and iron A. Miller (*Phys. Rev.*, 1959, **116**, Dec. 15, 1481–1482).

Magnetic properties and grain structure of Mn–Zn ferrites W. Heister (*J. Appl. Phys.*, 1959, **30**, April, 225–245) Relations between sintering temp., grain size, and magnetic properties leading to very low-loss materials are reported.

Metal–ceramic mixtures P. Murray (*Powder Met.*, 1960, (5), 64–80) A review of the properties and possibilities of oxide and carbide cermets.

The question of the structure stability of cermet materials on a basis of titanium carbide B. Čech (*Silikáty*, 1960, **4**, (1), 1–9).

PROPERTIES AND TESTS

Multichannel photoelectric scanning instrument for sizing microscopic particles B. B. Morgan and E. W. Meyer (*J. Sci. Instr.*, 1959, **36**, Dec., 492–501) Coal particles dispersed on a microscope slide are counted down to 1·5 μ agreeing within 15% with visual counts. Sizes of Al_2O_3 particles (5–20 μ) were also determined, and numbers of 1–5 μ particles on a thermal precipitator slide.

An apparatus for the rapid determination of grain size and structural analysis of fine-grained preparations H. F. Fischmeister and M. E. Möller (*Z. Metallk.*, 1959, **50**, Aug., 478–480) An instrument is described, with the help of which grain size measurements and grain structure analysis can be carried out. Measurements are carried out by a photoelectric pulse transmitter. The equipment can be used with reflecting and ordinary microscopes.

Some problems we meet L. H. Dawtrey (*Inst. Mech. Eng. Autom. Div., preprint*, 1959, Oct., 3–19) Metallurgical problems of automobile design and various testing methods for fatigue, impact, and service simulation of various components are considered.

The physical metallurgy of 12% chromium steels K. J. Irvine *et al.* (*JISI*, 1960, **195**, Aug., 386–405) [This issue].

Effect of structure on mechanical properties and ductility of aluminium-killed steel V. M. Chirkin and B. K. Barzii (*Stal'*, 1960, (1), 74–77) Experimental batches of rimming steel melted in a 200-t OH furnace and deoxidized with Al were rolled with or without water spray cooling and annealed. The cooled sheets showed a pancake-like grain structure while the uncooled sheets had the normal equiaxial ferrite grain structure. The effect is explained by the state of the AlN. On pressing, the former type with flattened grains was markedly superior and the process is recommended.

Metallurgy of mine roof supports (*Nat. Coal Board Sci. Bull.*, 1958, Autumn, (7), 14–19) Roof bar strength, brittle fracture, metal structure, and testing are briefly reviewed.

Fulmer Research Institute (*Met. Treatment*, 1959, **26**, Oct., 381–389) A description is given of the current investigations and equipment available in the establishment.—A.D.H.

Dept. of mining and metallurgical engineering T. A. Read (*Univ. Illinois Bull.*, 1959, **56**, June, (72), 73–83) A review of engineering research at the University 1957–1958. Diffusion in metals, anelasticity in metals and alloys, radiation damage, alloying behaviour of transition elements, diffusionless phase transformations, interface and film studies, criteria for cathodic protection, energy absorption of tempered martensitic structure in base and weld metal, hydrogen embrittlement studies of steel under cathodic protection, heat-treatment of ductile cast iron, effect of heat treatment and chemistry on properties of low density iron base sinterings, fatigue strength of pearlitic malleable iron and other topics of research are briefly outlined.

Standard for inspecting stern frames of cast steel (tentative) Y. Yokoyama (*Imono*, 1959, **31**, Oct., 1025–1030) [In Japanese].

Putting together of the results of Belgian researches on the correlation between size and physical properties of grey cast iron A. de Sy and J. van Eeghem (*Fond. Belge*, 1960, **30**, Feb., 38–43) Large and small bars were cast by methods shown. Hardness was correlated with various other factors and also tensile properties.

Developments in research techniques used in physical metallurgy in the United Kingdom G. A. Geach (*Jernkont. Ann.*, 1959, **143**, (12), 789–809) A lecture given at the meeting of Svenska Metallografförbundet on 22 May 1959, and dealing with the following topics: work at the temperature of liquid He, creep under constant shear stress, hardness testing at high temperatures, direct observation of dislocations, X-ray microanalysis, and applications of arc and electron bombardment heating.

Methods of determining surface roughness M. K. Testerman (*WADC Techn. Rep.*, 58–230; *AD*–155828; *PB* 151268, 1958, June, pp.42; from *US Techn. Rep.*, 1959, **31**, March 13, 166) Twelve methods are mentioned and the capacitance method is examined. Measurements on specimens coated with monomolecular films were not reproducible but polarization capacitance is aqueous solution was. Steel and other metals gave consistent values by the bridge method. Formamide and vinyl cyanide were not satisfactory. Preliminary studies on a tracer method were made.

Roughness measurement on cold-worked surfaces W. Lueg and U. Krause (*Stahl Eisen*, 1960, **80**, March 17, 325–336) The roughness of cold-worked, i.e. rolled and drawn, surfaces was measured by stylus instruments and the results are presented in roughness-distribution curves. The difference in the appearance of the curves for the two types of surface is discussed and the reasons for it explained. Difficulties in obtaining identical values in roughness measurement when using apparatus made by different manufacturers on different principles are pointed out and attempts are reported to induce the manufacturers to build instruments that show results of the same magnitude when used on the same surface. It was also found that photometric methods for a determination of surface roughness are unsuitable. The paper is followed by a long and lively discussion on the principles of surface-roughness measurement (22 refs).

Modulus of elasticity and damping capacity of grey cast iron E. Plénard (*Gjuteriet*, 1959, **49**, (8), 207–209) Methods of determining the modulus are outlined, and the effect of graphite structure on the modulus and damping capacity are discussed. (The article is based on the publication by the author: 'Considerations on the modulus of elasticity and damping capacity of grey cast iron', Centre Technique des Industries de la Fonderie, Paris).

The evaluation of internal stresses E. Houdremont and H. Scholl (*Z. Metallk.*, 1959, **50**, Sept., 503–511) The generation of first and second type internal stresses is described, and methods of determining stresses are reviewed and examples are given to show the effects on physical and mechanical properties. Although a condition of internal stress cannot completely and satisfactorily be determined by non-destructive methods, the influence of certain stresses can often be estimated theoretically.

Plastic deformation and the atomic lattice J. Orland (*Met. Elect.*, 1956, **20**, Aug., 45–56; Sept., 53–62; Oct., 46–55) [In Spanish] Modern theories of plastic deformation are reviewed (14 refs).—P.S.

Internal stresses in quenched steels H. Borchers and A. Safferling (*Stahl Eisen*, 1960, **80**, April 14, 541–546) This is a résumé of the thesis by the second author. The internal stresses present in samples of steel after quenching can be attributed to thermal stresses and to stresses as a result of phase transformations taking place during quenching. The effect of these stresses are discussed in the light of recent publications, and the reasons given why quenching in a salt bath appreciably reduces internal stresses.—T.G.

A replica technique for measuring static strains V. M. Hickson (*J. Mech. Eng. Sci.*, 1959, **1**, Sept., 171–183) A method measuring displacement of surface scratches is described.

Stress anisotropy in Ni–Fe thin films J. D. Blades (*J. Appl. Phys.*, 1959, **30**, April, 2605–2615).

Investigations on the validity of an idealized theory of plastic flow for rolled mild steels C. H. Avery, T. E. Tietz, and J. E. Dorn (*ATI*–36854; *PB* 139622, 1947, Dec., pp.55; from *US Res. Rep.*, 1959, **31**, June 12, 469) Tensile and bulging tests showed the steel to be anisotropic which appeared to be planar as the curves were practically identical at 0°, 45°, and 90° to the rolling direction. The assumption that generalized flow stress is a function of plastic strain energy for complete three-dimensional isotropy gave fair agreement.

Numerical value of the uniform and the concentrated plasticity of steel under tension V. I. Egiz (*Zavods. Lab.*, 1959, (12), 1486–1491) Under tension, plastic deformation is made up of two main factors, the uniform flow and the concentrated deformation. Hence total deformation depends on the values and ratios of these two components. In a large number of experiments a preferential connexion has been discovered between relative elongation and reduction of area, corresponding to the macro- and micro-structure of steel. It is also pointed out that the elucidation of plastic characteristics depends to a large extent on the approach of the investigator to the problems, for example, if he has to study the composition of an alloy, methods of production and the behaviour of the metal under tests for fracture, the chief role is played by the macrostructure and therefore the principal index will be given by the reduction of area.

Dimensional springback of steel bars after

drawing and pushing-in W. Lueg and O. Pawelski (*Stahl Eisen*, 1960, **80,** March 17, 343–350) The authors review the mechanisms involved in the dimensional springback of steel bars after drawing and pushing-in and report their own experiments on the subject in which they found that the formulae used for the calculation of the dimensional changes generally give values that are too large. They propose a new formula which gives more realistic values under certain conditions which are stated. The relation between dimensional springback and residual stresses is briefly studied.—T.G.

National and international aspects of stress analysis M. L. Meyer (*Engineer*, 1960, **209.** Jan. 15, 90–93) An account of the Delft conference. Bodies in various countries engaged on research are referred to and items discussed are briefly reviewed and illustrated.

Damping capacity—its measurement and significance J. W. Jensen (*US Bur. Mines, Rept. Invest.*, 5441, 1959, pp.46) Theory, measurement, instrumentation, and interpretation are reviewed (160 refs).

The nature of internal friction in quenched and annealed steel M. A. Krishtal and S. A. Golovin (*Fiz. Met. Metalloved.*, 1959, (2), 294–301) Quenched and annealed steels with 0·71 and 0·92%C were investigated. Maximum friction is noticed between 228 and 255°. It falls with increasing annealing temp. and diminishing carbon content. The peaks increase nearly linearly with increasing residual austenite in the quenched steel, also with increasing carbon content. Calculation of the diffusion coefficients and of the activation energy have proved that internal friction is related to the carbon diffusion in the residual austenite.

The nature of the relative damping of oscillations in quenched and low annealed steels M. A. Krishtal and S. A. Golovin (*Fiz. Met. Metalloved.*, 1959, (2), 302–308) Investigations were carried out on 50A and U7A steels, respectively 0·49 and 0·70%C after single and two-fold quenching, and after heat treatment at −196° and at higher temp. These have shown that the relative damping is related to the presence of the residual austenite and its decay products during the anneal. With a large C content (U7A) under high stress, the relative damping of oscillations is reduced. A second quenching diminishes the property of these steels to extinguish these vibrations.

Effect of internal deformation on the internal friction of the iron–tungsten alloy V. S. Postnikov and R. S. Lebedev (*Fiz. Met. Metalloved.*, 1959, (1), 95–102) The experimental results confirm the conclusion that internal friction at high temp. may serve as a special indication of the strength of the alloy. The alloy is the more heat-resistant the less its internal friction. This condition is inescapable as has been shown by many investigators, but it is not yet sufficiently understood so that its explanation is not yet given.

Effect of plastic deformation on the internal friction of iron and of iron–nickel alloy R. S. Lebedev and V. S. Postnikov (*Fiz. Met. Metalloved.*, 1959, (2), 310–314) Investigations were carried out as described previously (preceding abstract) but at a high rate of heating, i.e. 6°/min. The results obtained with electrolytic iron after reductions of 8; 17; 30; 47; 70; and 92%, and on Fe–Ni alloy (Armco + 4%Ni) reduced between 20 to 80% are tabulated. It may be deduced from these data that the reduction in the internal friction in pure metals and alloys is not due to diffusion but to other causes. Regarding the dislocation mechanism of the process, there are not yet sufficient data available to justify definite conclusions.

The effect of deformation on the internal friction of iron measured at the carbon peak position F. W. C. Boswell (*Canad. J. Phys.*, 1959, **37,** Dec., 1474–1481) A time law is followed predictable from carbon segregation to dislocations. The amount of C in random solid solution is increased by deformation, some derived from carbide.

Methods and results of the measurement of internal friction in steels and non-ferrous metals H. J. Seemann (*Inst. Hierro Acero*, 1959, **12.** Oct.–Dec., 295–309) In a review of the results of investigations of internal friction of metals, the Snoek experiments are described in detail. Experimental methods developed in the author's laboratories are described and results obtained in tests on Al and Al alloys are discussed (19 refs).—P.S.

Temperature relation between internal friction and the elastic constants of pure iron V. M. Amonenko, B. I. Shapoval, and V. V. Lebedev (*Fiz. Met. Metalloved.*, 1959, (2), 249–254) The relation for pure iron (99·99%) is similar to that for other metals but the absolute values for all temp. investigated are higher by a small factor than for Armco iron. Not all the carbon contained in the pure iron is present in the solid solution. Between 20 and 700° there exists an almost linear relation between the modulus of elasticity and the shear modulus.

The effect of impurities on the electron density of metals, II E. Daniel (*J. Phys. Radium*, 1959, **20,** Nov., 849–859).

A few topics of dislocation A. Ookawa (*Metal Phys.*, 1959, **5,** Nov., 223–231) [In Japanese] A review with captions in English.

Dislocation-locking by carbon, nitrogen and boron in α-iron I. Codd and N. J. Petch (*Phil. Mag.*, 1960, **5,** Jan., 30–42) Pure iron was heated in wet H_2 till C and N were removed as shown by yield-point suppression and C was replaced giving nitrogen-free steels with 0·2–0·26 and 0·09–0·12%C and 0·47%Mn. Replacement of 0·008 and 0·013%N in the carburized steels was also carried out. The locking by C and N was compared and also B was used. Boron locking is stronger than that of N_2 which in turn, unexpectedly, is greater than that of C. The ductile-brittle transition temp. is raised by the stronger locking.

Calculation of the cohesive energy of metallic iron F. Stern (*Phys. Rev.*, 1959, **116.** Dec. 15, 1399–1417) The calculation is made for the body-centred cubic structure in the singlet spin state at 0°K.

The vapour pressure of inorganic substances, VII: Iron between 1356°K and 1519°K and cobalt between 1363°K and 1522°K J. W. Edwards, H. L. Johnston, and W. E. Ditmars (*TR*–281–9; *ATI*–107836; *PB* 143212, 1951, March 21, pp.16; from *US Res. Rep.*, 1959, **32,** Dec. 11, 683) Heats of sublimation are calculated. The accommodation coefficient is unity.

Vacuum distillation of metals with their condensation on heated surfaces V. M. Amonenko, B. M. Vasyutinskii, V. V. Lebedev, and B. I. Shapoval (*Fiz. Met. Metalloved.*, 1959, (6), 862–867) By the vacuum distillation described and by the condensation on a heated surface it is possible to obtain, after a first distillation, an iron of 99·98% purity. A second distillation reduces the total impurities to 0·01%. Fe–Mn can be obtained 99·96% pure with a yield of about 80% of the metal contained in the alloy.

The influence of additions on the activity of carbon in liquid iron alloys H. Schenck (*Rev. Mét.*, 1960, **57,** Jan., 3–11) The influence of Si on the activity of C in low-C alloys is examined, and the investigation is extended to high-C alloys. The coefficient of activity of C and the coefficients of interaction of a large number of alloying constituents with the variation of C concentration at saturation are determined, and the relations between these coefficients of interaction and the position of the elements in the periodic classification is shown.

The solid solubilities of iron and nickel in beryllium S. H. Gelles (*NMI*–1207, 1958, July, pp.35; from *US Res. Rep.*, 1959, **31,** April 17, 273; no abstract).

The lattice constants of the body-centred cubic phases FeTi, CoTi, and NiTi H. P. Stüwe and Y. Shimomura (*Z. Metallk.*, 1960, **51,** March, 180–181) The lattice constants in the system Ti with Fe, Co, and Ni were determined up to the intermetallic compounds FeTi, CoTi, and NiTi, after quenching from 1000°C. The lattice constants and their change with composition are given in a diagram.—T.G.

Structure of the energy spectrum of the electrons of the Fe–Cr and Fe–Cr–Ni alloys N. D. Borisov and V. V. Nemoshkalenko (*Fiz. Met. Metalloved.*, 1959, (2), 211–215) This work studies the changes in the structure of the energy spectrum of the electrons of the 'external' strips of Fe–Cr alloys by the partial changes in the percentage of the alloy component and of Ni, and in temp. The data are tabulated.

Thermodynamics and formulation of reactions involving imperfections in solids F. A. Kröger, F. H. Stieltjes, and H. J. Vink (*Philips Res. Rep.*, 1959, **14,** Dec., 557–601) The Schottky system of amount-of-substance variables refers to building units and can be associated with thermodynamic potentials and phase rule components. The system of Wagner and others refers to structure elements and thermodynamic potentials cannot be assigned; however by introducing virtual potentials they allow a treatment often more convenient than the former. Various types of disorder are formulated.

On topotaxy J. D. Bernal (*Schweiz. Arch.*, 1960, **26,** Feb., 69–75) [In English] The transformation of one crystal form into another of different composition without a large alteration of the axial direction has been called topotaxy by Gorter. It is claimed that topotaxy is very widespread and may even be the most common mechanism of transformation in the solid state. The structural transformations involved in rusting, i.e. the iron oxides and hydrates, and the ferrite transformations are discussed at some length (19 refs).—T.G.

New means of tensile testing A. G. Epprecht (*Schweiz. Arch.*, 1960, **26,** Feb., 55–59) The author describes the Drage-Multitest tensile testing machine in which the deformation speed is load-independent and constant. This apparatus is used in conjunction with an extensometer which directly plots a stress/strain curve. A feature of this little instrument is that the stress/strain curve can be plotted at various amplifications, e.g. at the beginning at 500× and later at 5×. Although primarily developed for the testing of plastics, it may also be used for testing metals, and as an example the stress/strain curve of a steel is shown.—T.G.

Design for an autographically recording tensile machine M. A. Adams (*J. Sci. Instr.*, 1959, **36,** Oct., 444–446) A design for dry-box or hot-cell work is described.

Elastic limit of AP shot steels as a function of tempering temperature C. M. Carman (*PB* 135532, 1957, Oct., pp.19; from *US Res. Rep.*, 1959, **31,** March 13, 164) The elastic limit of 98 V65, 4160, and Mn–Mo armour-piercing shot steels was measured. The two former show a max. after tempering at 500–700°F and the latter at 800°F.

Tensile tester of revolutionary design developed for the wire industry D. C. Scott jun. (*Wire Wire Prod.*, 1959, **34,** June, 704–705) The 'CRE' tensile tester is described; it has been designed for routine testing and research work on bare and insulated wire, and cable components. Precise yet simple stress-strain evaluations are made possible by the inertialess, electric weighing, constant-rate-of-extension method used.—D.L.C.P.

Application of induction heating to short-time elevated temperature tensile testing A. P. Levitt and A. G. Martin (*PB* 151848; 1959, June, pp.16) The equipment and operation are described. The application is non-ferrous.

Serrations in the stress/strain curve of cold-worked 301 stainless steel at 20°K J. F. Watson and J. L. Christian (*JISI*, 1960, **195,** Aug., 439) [This issue].

Destructive testing for quality control in weldments R. E. Lismer (*Brit. Weld. J.*, 1959, **6,** Oct., 439–442) The tensile, bend, and impact test methods are briefly discussed. More stress is laid on the right choice of electrodes and the necessary analysis control of electrodes and their deposits.—U.E.

Strength of metals; suggested classifications for steels and other alloys G. F. Lee (*Iron Steel*, 1960, **33,** Jan., 9–10) Classification based on proof stress and form of the material—whether as-cast or in finished or semi-finished state, etc. is considered.

Effect of specimen taper on determination of elongation in the tension test E. B. Kula and F. R. Larson (*ASTM Bulletin*, 1959, (238), May, 58–61) It is shown that the 1% difference in diameter permitted in the ASTM standards for round tensile specimens can lead to differences in local strain of > 30%.

The notch ductility and tensile properties of some synthetic mild steels A. M. Sage and

F. E. L. Copley (*JISI*, 1960, **195**, Aug., 422–438) [This issue].

The experimental determination of the stress conditions in the adhesives layer of overlapped metal junctions A. Matting and K. F. Hahn (*Z. Metallk.*, 1959, **50**, Sept., 528–533) Experiments on six metal adhesives of different composition showed that the stress-strain characteristic is non-linear, and that results calculated on the assumption that Hooke's law is obeyed are too high. Stress distribution in the adhesive layer of overlapping junctions is examined, and it is shown that the creep process is influenced by relaxation in fatigue-tested specimens.

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Evaluating high-strength steels under complex loading E. A. Loria (*Mach. Design.*, 1960, **32**, Feb. 4, 120–125) Notch tests and burst tests are discussed and net-fracture stress is defined and interpreted. The importance of homogeneity in design and the avoidance of stress-concentrating features or defects is stressed.

The effects of inelastic action on the resistance to various types of loads of ductile members made from various classes of metals. Part IX. T-section eccentrically-loaded tension members made of type 304 stainless steel and tested at 1000°F O. M. Sidebottom and S. Dharmarajan (*WADC Tech. Rep.*, 56–330; *ASTIA Doc.* 202496; *PB* 151674, 1958, Oct., pp.12) A theoretical and experimental study of the load necessary to produce a specified amount of inelastic deformation on this material when loaded eccentrically at 1000°F gave good agreement between predicted and measured results, provided that initial eccentricity was small enough to ensure that yielding did not occur in the most highly stressed compression fibres. With larger eccentricity, experimental curves fell appreciably below theoretical results over the range where yielding had taken place on the compression side of the member. (*ASTIA Doc.* No.241767; *PB* 151894, 1959, May, pp.28) Eccentrically-loaded T-sections of 17–7 PH stainless steel were loaded at 1000°F. The deflection of the central section of each member was accurately predicted by both arc hyperbolic sine theory and interaction-curve-moment-load curve theory. Tension creep data were also obtained. An error in Part IX is pointed out.

Torsion tests on galvanized wire O. F. Holm (*Wire*, 1959, June, 69–70) Factors influencing the torsion strength of hot-dip galvanized wire are discussed with reference to the pretreatment and the galvanizing process. It is concluded that a satisfactory indication of the mechanical properties of the wire can only be obtained by previously pickling and stripping the wire.

Strength of thick cylinders subjected to repeated internal pressure J. L. M. Morrison, B. Crossland, and J. S. C. Parry (*Inst. Mech. Eng., preprint*, 1959, Oct., pp.16) An extension of the previous work on fatigue under triaxial stress reported in 1956. Work on 2½%Ni–Cr–Mo steel tested up to 10^7 repetitions of oil pressure up to 20 t/in² is described.

Impact testing machines J. H. Lamble (*Engineering*, 1959, **188**, Oct. 16, 352) Izod and Charpy tests were carried out with the machines on various different floors or supports. Differences in Charpy values were small but in Izod values they were considerable.

A remotely controlled drop-weight test machine for brittle-fracture studies L. E. Steele and J. R. Hawthorne (*NRL*–5278, 1959, Jan., pp.8; from *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1205) For testing large pieces for embrittlement due to radiation a notch-impact device was constructed. Thermally-conditioned insulated chambers are used.

Effect of specimen geometry on Charpy low-blow transition temperature G. M. Orner (*WAL TR* 112/87–6; *PB* 140642, 1953, July, pp.21; from *US Res. Rep.*, 1959, **32**, Aug. 14, 239).

Effects of alloying elements on impact properties of quenched and tempered steels M. Baeyertz, W. F. Craig, jun., and J. P. Sheehan (*ATI*–66620; *PB* 136745, 1949, Sept., pp.320; from *US Res. Rep.*, 1959, **32**, Aug. 14, 238) Alloys containing Mn, Ni, Cr, or combinations, with or without Mo were tested at −320°F to a temp. above the transition range.

Investigation of low-temperature toughness n ordnance steel C. Wells and W. B. Triplett (*WAL* 310/204–1; *PB* 135323, 1957, Oct., pp.101; from *US Res. Rep.*, 1959, **31**, April 17, 249) Non-metallic inclusions and segregation are the major causes of impact-value variations. Three special ingots were prepared, the one treated with rare earths was superior in toughness, that treated with Ca–Si was second and the third with Zr–Si had least toughness.

Charpy-V transition temperatures of ferritic iron alloys: Part II. Iron–chromium and iron–chromium–aluminium alloys J. E. Srawley (*NRL Rep.* 5367; *OTS* 151811, 1959, Aug. 24, pp.15) Alloys up to 21·3%Cr and three series of Cr–Al alloys with up to 24%Cr and 6%Al were examined. The type of fracture below transition temp. was transgranular and the temp. decreased with increasing Cr when quenched from 1650°C and increased when slowly cooled. Alloys with ~0·5%Al had lower and those with ~5%Al had higher transition temp. than without Al. Mn up to 12·8% had no beneficial effect and Ti up to 1·92% raised transition temp.

Metallurgical examination of a failed electromagnetic liquid metal pump channel G. H. Broomfield (*Metallurgia*, 1959, **60**, Nov., 201–203) Failure by fatigue of a 13%Cr steel liquid metal pump channel was probably due to pressure pulses in the liquid metal, with evidence that liquid metal corrosion had some effect in causing the failure. (17 refs)

Application of the Locati method of using an individual part for controlling the fatigue limit of various automobile parts C. Daubertis, J. Beuret, M. Renout, and R. Cazaud (*Met. Ital.*, 1959, **51**, Dec., 541–547) In an earlier paper the authors proved the validity of the Locati method for the inspection of mass-produced surface hardened wheel hubs. The present paper discusses the extension of this method to other parts—rocker arms, rods, etc., and in the case of wheel hubs using statistical tests. The results obtained confirm the usefulness of the Locati method for the control of mass-produced parts. According to statistical tests carried out on wheel hubs the fatigue limit shows no appreciable difference from that found by established long time tests at 5×10^6 cycles and the spread of results is of the same magnitude in all cases.

Leaded steels: Apt to fatigue? J. R. Shure and G. W. Brock (*Iron Age*, 1959, **184**, Dec. 3, 120–121) Test results are reported on the effect of Pb additions on the fatigue properties of AISI 4340 steel. Up to 0·25%Pb lowered the

mean fatigue limit but had little effect on fatigue life at stresses above the mean fatigue limit.—A.G.

Effect of residual stresses on the fatigue resistance and on the service life of forged cold-rolling rolls P. Funke (*Stahl Eisen*, 1960, **80,** March 17, 368–370) A review of several recent Russian publications on the factors that affect the service life of cold-rolling rolls. The Russian studies elucidate the concept of the correlations between working stresses imposed on the rolls and the residual stresses which are a result of the mode of fabrication.—T.G.

A relation between the critical alternating propagation stress and crack length for mild steel N. E. Frost (*Inst. Mech. Eng. preprint*, 1959, pp.19) Fatigue tests on specimens with edge cracks are reported and the alternating stress just sufficient to cause crack growth was determined. A relationship for critical stress to crack length is found and a wide range of notch fatigue data is correlated.

Effect of hydrogen on the fatigue resistance of welded steel W. Hofmann and G. Vibrans (*Schweissen Schneiden*, 1960, **12,** March, 95–102) The effect of H_2 on the fatigue resistance of two low-alloy steels was studied. It was found that the steels after being loaded with H_2, lost half of it in the first 20 h at room temp. the fatigue studies were therefore also conducted at −45°C. Fatigue resistance is not decreased by H_2 unless the steel has been tensile stressed by 10% before fatigue testing. In this case the damage done is permanent, even if all the H_2 has been driven off. This behaviour is the same with welds. In welds, however, another factor may contribute to reduce fatigue resistance; microcracks occur as a result of too high cooling rates when passing through the temp. range around 200°C (29 refs).—T.G.

A general approach to the practical solution of creep problems A. Mendelson, M. H. Hirshberg, and S. S. Manson (*Trans. ASME, J. Basic Eng.*, 1959, **81D,** Dec., 585–598) The method described solves creep problems by use of successive approximations, and is valid for different creep laws and loads. Examples are given of creep in a flat plate with a temp. gradient, and in rotating discs, with and without the inclusion of transient conditions.

Experimental work on malleable iron L. C. Marshall, G. Sommer, and D. A. Pearson (*Mod. Cast.*, 1960, **37,** Jan., 109) The results obtained so far in a series of tests on the stress-rupture and creep properties of ferritic and pearlitic malleable irons are summarized. These tests were carried out at 800, 1000, and 1200°F. The object of the programme is to establish tentative use specifications for these irons when used for prolonged periods at elevated temp.

Nonlinear creep deformations of columns of rectangular cross section H. H. Bleich and O. W. Dillon, jun. (*J. Appl. Mech.*, 1959, **26,** Dec., 517–525) Mathematical. The law adopted is $\dot{\epsilon}=(1-E)\dot{\sigma}+\lambda\sigma^n$.

A phenomenological theory for the transient creep of metals at elevated temperatures E. Z. Stowell (*NACA TN* 4396, 1958, Sept., pp.31; from *US Res. Rep.*, 1959, **31,** March 13, 167) An alloy of two metallic phases, each with its own elasticity and viscosity, will exhibit transient creep after application of a constant stress. Four metals show entire families of creep curves to be given by a single set of constants.

Creep-rupture strength of austenitic Cr–Ni–Mo steels in sheet and bar form G. V. Smith, F. Garofalo, R. W. Whitmore, and R. R. Burt (*Trans. ASME, J. Basic Eng.*, 1959, **81D,** Dec., 610–616) It is shown that small variations in C, N, and P in 18%Cr–8%Ni–Mo bar and sheet account for about ½ of the 100 and 1000 h creep-rupture strength in tests at 1100, 1300, and 1500°F.

Correlation of high temperature creep and rupture data H. Conrad (*Trans. ASME J. Basic Eng.*, 1959, **81D,** Dec., 617–628) The problem of correlating creep and rupture data at $T>0{\cdot}45T_m$ is examined and it is concluded that deformation of metals and alloys is diffusion controlled at these temp.; activation energy is given by that for the self-diffusion of the base metal or alloying constituents, and should be independent of stress. An equation is derived for creep rupture having the form: $\dot{\epsilon}_s=\frac{K}{t_r}\approx A\exp-\left(\frac{\Delta H_D}{RT}\right)\sinh\left[\sigma/\sigma_0(T)\right]^n$. Where $\dot{\epsilon}_s$ is minimum creep rate, t_r the time to rupture ΔH_D a diffusion activation energy, σ the stress, and K, A, $\sigma_0(T)$, and n are constants. Data for several superalloys and an Al alloy were found to be in agreement with this equation (55 refs).

The creep of strain gauges as a rheological problem C. Rohrbach and N. Czaika (*Materialprüf.*, 1960, **2,** March, 98–105) The creep of strain gauges and its probable origin is discussed. The creep is attributed to a yielding of the cement as a result of the restoring forces of the strained wire. The authors give a diagram from which the creep of the strain gauge may be deduced. They also discuss the question to what extent strain gauges may be used for measurement of creep.—T.G.

Creep-resisting materials S. Arwidson (*Jernkont. Ann.*, 1959, **143,** (7), 373–413) Creep resisting materials are enumerated, and their properties and applications described. The materials dealt with are: martensitic steels, austenitic steels, and Ni- and Co-base alloys, with a section on cements (11 refs).

A method of treating the results of long-term strength tests V. I. Nikitin (*Zavods. Lab.*, 1959, (12), 1492–1496) The article presents a simpler and more accurate method of determining the breaking load for any given time, based on F. R. Larson and J. Miller's parameter (*Trans. ASME* 74, (5), 1952) $T(C+\log\tau)=p(\delta)$ where T is the absolute temperature; τ, the time to fracture, C constant, and δ the load. In this equation the relation between the product T log τ and temp. at constant load is linear and the tangent of the slope of such a line to the temp. axis is always equal to C. The graph T log τ −T is therefore made up of a family of straight lines, each of which corresponds to a value of the load. This gives the possibility of constructing a convenient diagram for long-term resistance of the material from which it is possible to determine the ultimate long-term strength for any period of time.

Progress reports of investigation of railroad rails R. E. Cramer (*Univ. Illinois Bull.*, 1959, *Reprint series No.*60, April, pp.13) Failures in control-cooled rails and shelly rail studies are included. A summary of 25 failures from 11 causes and of shelly failures in rolling-load tests is presented.

Progress report of investigation of railroad rails and joint bars R. E. Cramer and R. S. Jensen (*Univ. Illinois Bull., Reprint series No.*59, 1958, May, pp.33) Failures in control-cooled rails, rolling-load tests of joint bars, shelly rail studies, and tests on continuous welded rails make up the report.

Some crankshaft failures: investigations, causes and remedies R. Atkinson and P. Jackson (*Shipbuilding Ship Rec.*, 1960, **95,** Feb. 11, 180–181) Failures of Diesel engine shafts are briefly reviewed and the methods of examination are given.

Third International Symposium on hardness measurement in theory and practice W. Bischof and K. Meyer (*Materialprüf.*, 1960, **2,** March, 107–114) The third international symposium on 'hardness measurement in theory and practice' was held at Dortmund on 23–25 Sept. 1959. The present report gives the titles and abstracts of 28 papers by contributors from all over the world. The papers and discussions will be published in full by the VDI, Düsseldorf in the near future.—T.G.

The technique of hardness testing V. L. James (*Sheet Met. Ind.*, 1960, **37,** Feb., 125–135) This final instalment refers to Cu and its alloys.

A simplified calculation of Brinell hardness R. Böklen (*VDIZ*, 1960, **102,** Feb. 1, 128) Brinell hardness H_B is determined from an equation $H_B-H_M=Kx$, where H_M is the Meyer hardness and Kx a corrective term depending solely on the load.

Hardness testing of thin sheet and tinned sheet E. Wondris and V. Seul (*Stahl Eisen*, 1960, **80,** March 17, 356–365) The authors discuss the advantages offered by using the super Rockwell hardness testing for thin sheet and tinned sheet and report on studies of the mechanism involved in the impression of thin sheet, they also discuss the effect of the thickness of the sheet and that of the support on the results. The correlation with other methods and the use of conversion factors are touched upon.—T.G.

Measuring scratch depth (*Ind. Fin.*, 1960, **12,** March, 57) Details are given of an instrument developed by Boeing Aircraft Co. for measuring the depth of scratches ranging from 0·030 to 0·00005 in. The instrument measures the length of the shadow cast by a beam of light at a pre-determined angle.—G.F.

Conversion chart for microhardness values R. A. Slepian (*Mat. Design Eng.*, 1960, **51,** Jan., 123) A graph designed for converting microhardness values to the more commonly used microhardness scales is illustrated.

Hardenability studies on 4340 steel E. R. Klier, V. Weiss, and G. Sachs (*PB* 130766, 1954, Oct., pp.46; from *US Govt. Res. Rep.*, 1959, **31,** June 12, 470) Jominy tests were carried out after various heat treatments. Metallographic constituents were determined and X-ray studies were made.

Some considerations on the problem of brittle fracture O. Werner (*Materialprüfung.*, 1959, **1,** June, 189–200) Modern theories of metal pasticity are discussed, and the contribution of dislocation theory to present knowledge of the causes of brittle fracture is reviewed. Experimental results in the field of internal friction, and the results of electron microscopical investigations are reported, which support theories based on dislocation mechanism, and in which dislocations can be viewed directly (43 refs).

Embrittlement of solid metals in a liquid metal W. A. Morgan (*Met. Treatment*, 1959, **26,** Sept., 333–339) A review of causes of embrittlement and cracking of steels and nonferrous metals (57 refs).

Micromechanism of brittle fracture in a low-carbon steel G. T. Hahn, W. S. Owen, B. L. Averbach, and M. Cohen (*Weld. J.*, 1959, **38,** Sept., 275s–276s) Tensile data, elastic limit measurements, and metallographic observations of slip, twinning, and micro-cracks, for samples tested in the range room temp. to −250°C, as well as standard Charpy data, are presented for a low-carbon steel in a coarse (ASTM GS No.4) and a fine-grained (ASTM GS No.7) condition.

Brittleness of cobalt-iron alloys C. W. Chen and G. W. Wiener (*J. Appl. Phys.*, 1959, **30,** April, 199s–201s) A mechanism is advanced.

On hydrogen brittleness of steel: III. Metallurgical observations I. Onishi and Y. Kikuta (*Techn. Rep. Osaka Univ.*, 1960, **10,** Jan., 193–208) [In English] The cracks and fracture surfaces formed in embrittled cathodically charged mild steel are described.

Hydrogen embrittlement of plated steel springs J. A. Gurklis, L. D. McGraw, and C. L. Faust (*PB* 151125, 1956, Dec., pp.46; from *US Res. Rep.*, 1959, **31,** April 17, 249) With reasonable care no subsequent heat-treatment or other treatment is necessary after Cd plating.

Brittleness of special steels in the as-cast condition: influence of impurities R. Castro and A. Gueussier (*Rev. Mét.*, 1960, **57,** Feb., 135–148) The influence of structure and composition on the brittleness of an Ni–Cr–Mo steel and an austenitic Ni–Cr steel in the as-cast condition were investigated by means of temp.-impact strength curves. Topics considered were: the effects of mould size, and hence the cooling rate on minor segregation and brittleness; influence of various impurities on intergranular brittleness in heat-treatable Ni–Cr–Mo steels; influence of a homogenizing thermal treatment on brittleness; and the effects of various impurities on the brittleness of as-cast austenitic stainless steels, together with the modifying influence on the results of the presence of 5–10% ferrite in the structure.

Experimental study of solidification cracks in steels: influence of impurities A. Gueussier and R. Castro (*Rev. Mét.*, 1960, **57,** Feb., 117–134) The experimental technique used by the authors is described in detail; one feature is the simultaneous preparation in a single mould of 16 specimens, to reduce the number of tests and the effects of dispersion. Results obtained on Ni–Cr austenitic steels show the influence of primary structure on crack sensitivity; com-

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Effect of shear deformations on the bending of rectangular plates V. L. Salerno and M. A Goldberg (*Trans. ASME J. Appl. Mech.*, 1960, **27E**, March, 54–58).

Carbon steel sheets for stamping and deep drawing (*Tecn. Indust.*, 1959, **37**, Dec., 1821–1825) Tests of ductility are described in detail. The Erichsen number is favoured most for determining the best sheet for each class of work.—P.S.

Evaluating high-strength steels under complex loading E. A. Loria (*Mach. Design.*, 1960, **32**, Feb. 4, 120–125) Notch tests and burst tests are discussed and net-fracture stress is defined and interpreted. The importance of homogeneity in design and the avoidance of stress-concentrating features or defects is stressed.

The effects of inelastic action on the resistance to various types of loads of ductile members made from various classes of metals. Part IX. T-section eccentrically-loaded tension members made of type 304 stainless steel and tested at 1000 F O. M. Sidebottom and S. Dharmarajan (*WADC Tech. Rep.*, 56–330; *ASTIA Doc.* 202496; *PB* 151674, 1958, Oct., pp.12) A theoretical and experimental study of the load necessary to produce a specified amount of inelastic deformation on this material when loaded eccentrically at 1000°F gave good agreement between predicted and measured results, provided that initial eccentricity was small enough to ensure that yielding did not occur in the most highly stressed compression fibres. With larger eccentricity, experimental curves fell appreciably below theoretical results over the range where yielding had taken place on the compression side of the member. (*ASTIA Doc.* No.241767; *PB* 151894, 1959, May, pp.28) Eccentrically-loaded T-sections of 17–7 PH stainless steel were loaded at 1000°F. The deflection of the central section of each member was accurately predicted by both arc hyperbolic sine theory and interaction-curve-moment-load curve theory. Tension creep data were also obtained. An error in Part IX is pointed out.

Torsion tests on galvanized wire O. F. Holm (*Wire*, 1959, June, 69–70) Factors influencing the torsion strength of hot-dip galvanized wire are discussed with reference to the pretreatment and the galvanizing process. It is concluded that a satisfactory indication of the mechanical properties of the wire can only be obtained by previously pickling and stripping the wire.

Strength of thick cylinders subjected to repeated internal pressure J. L. M. Morrison, B. Crossland, and J. S. C. Parry (*Inst. Mech. Eng., preprint*, 1959, Oct., pp.16) An extension of the previous work on fatigue under triaxial stress reported in 1956. Work on 2¼%Ni–Cr–Mo steel tested up to 10^7 repetitions of oil pressure up to 20 t/in² is described.

Impact testing machines J. H. Lamble (*Engineering*, 1959, **188**, Oct. 16, 352) Izod and Charpy tests were carried out with the machines on various different floors or supports. Differences in Charpy values were small but in Izod values they were considerable.

A remotely controlled drop-weight test machine for brittle-fracture studies L. E. Steele and J. R. Hawthorne (*NRL*–5278, 1959, Jan., pp.8; from *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1205) For testing large pieces for embrittlement due to radiation a notch-impact device was constructed. Thermally-conditioned insulated chambers are used.

Effect of specimen geometry on Charpy low-blow transition temperature G. M. Orner (*WAL TR* 112/87–6; *PB* 140642, 1953, July, pp.21; from *US Res. Rep.*, 1959, **32**, Aug. 14, 239).

Effects of alloying elements on impact properties of quenched and tempered steels M. Baeyertz, W. F. Craig, jun., and J. P. Sheehan (*ATI*–66620; *PB* 136745, 1949, Sept., pp.320; from *US Res. Rep.*, 1959, **32**, Aug. 14, 238) Alloys containing Mn, Ni, Cr, or combinations, with or without Mo were tested at −320°F to a temp. above the transition range.

Investigation of low-temperature toughness n ordnance steel C. Wells and W. B. Triplett (*WAL* 310/204–1; *PB* 135323, 1957, Oct., pp.101; from *US Res. Rep.*, 1959, **31**, April 17, 249) Non-metallic inclusions and segregation are the major causes of impact-value variations. Three special ingots were prepared, the one treated with rare earths was superior in toughness, that treated with Ca–Si was second and the third with Zr–Si had least toughness.

Charpy-V transition temperatures of ferritic iron alloys: Part II. Iron–chromium and iron–chromium–aluminium alloys J. E. Srawley (*NRL Rep.* 5367; *OTS* 151811, 1959, Aug. 24, pp.15) Alloys up to 21·3%Cr and three series of Cr–Al alloys with up to 24%Cr and 6%Al were examined. The type of fracture below transition temp. was transgranular and the temp. decreased with increasing Cr when quenched from 1650°C and increased when slowly cooled. Alloys with ~0·5%Al had lower and those with ~5%Al had higher transition temp. than without Al. Mn up to 12·8% had no beneficial effect and Ti up to 1·92% raised transition temp.

Metallurgical examination of a failed electromagnetic liquid metal pump channel G. H. Broomfield (*Metallurgia*, 1959, **60**, Nov., 201–203) Failure by fatigue of a 13%Cr steel liquid metal pump channel was probably due to pressure pulses in the liquid metal, with evidence that liquid metal corrosion had some effect in causing the failure. (17 refs)

Application of the Locati method of using an individual part for controlling the fatigue limit of various automobile parts C. Daubertis, J. Beuret, M. Renout, and R. Cazaud (*Met. Ital.*, 1959, **51**, Dec., 541–547) In an earlier paper the authors proved the validity of the Locati method for the inspection of mass-produced surface hardened wheel hubs. The present paper discusses the extension of this method to other parts—rocker arms, rods, etc., and in the case of wheel hubs using statistical tests. The results obtained confirm the usefulness of the Locati method for the control of mass-produced parts. According to statistical tests carried out on wheel hubs the fatigue limit shows no appreciable difference from that found by established long time tests at 5×10^6 cycles and the spread of results is of the same magnitude in all cases.

Leaded steels: Apt to fatigue? J. R. Shure and G. W. Brock (*Iron Age*, 1959, **184**, Dec. 3, 120–121) Test results are reported on the effect of Pb additions on the fatigue properties of AISI 4340 steel. Up to 0·25%Pb lowered the

mean fatigue limit but had little effect on fatigue life at stresses above the mean fatigue limit.—A.G.

Effect of residual stresses on the fatigue resistance and on the service life of forged cold-rolling rolls P. Funke (*Stahl Eisen*, 1960, **80,** March 17, 368–370) A review of several recent Russian publications on the factors that affect the service life of cold-rolling rolls. The Russian studies elucidate the concept of the correlations between working stresses imposed on the rolls and the residual stresses which are a result of the mode of fabrication.—T.G.

A relation between the critical alternating propagation stress and crack length for mild steel N. E. Frost (*Inst. Mech. Eng. preprint*, 1959, pp.19) Fatigue tests on specimens with edge cracks are reported and the alternating stress just sufficient to cause crack growth was determined. A relationship for critical stress to crack length is found and a wide range of notch fatigue data is correlated.

Effect of hydrogen on the fatigue resistance of welded steel W. Hofmann and G. Vibrans (*Schweissen Schneiden*, 1960, **12,** March, 95–102) The effect of H_2 on the fatigue resistance of two low-alloy steels was studied. It was found that the steels after being loaded with H_2, lost half of it in the first 20 h at room temp. the fatigue studies were therefore also conducted at −45°C. Fatigue resistance is not decreased by H_2 unless the steel has been tensile stressed by 10% before fatigue testing. In this case the damage done is permanent, even if all the H_2 has been driven off. This behaviour is the same with welds. In welds, however, another factor may contribute to reduce fatigue resistance; microcracks occur as a result of too high cooling rates when passing through the temp. range around 200°C (29 refs).—T.G.

A general approach to the practical solution of creep problems A. Mendelson, M. H. Hirshberg, and S. S. Manson (*Trans. ASME, J. Basic Eng.*, 1959, **81D,** Dec., 585–598) The method described solves creep problems by use of successive approximations, and is valid for different creep laws and loads. Examples are given of creep in a flat plate with a temp. gradient, and in rotating discs, with and without the inclusion of transient conditions.

Experimental work on malleable iron L. C. Marshall, G. Sommer, and D. A. Pearson (*Mod. Cast.*, 1960, **37,** Jan., 109) The results obtained so far in a series of tests on the stress-rupture and creep properties of ferritic and pearlitic malleable irons are summarized. These tests were carried out at 800, 1000, and 1200°F. The object of the programme is to establish tentative use specifications for these irons when used for prolonged periods at elevated temp.

Nonlinear creep deformations of columns of rectangular cross section H. H. Bleich and O. W. Dillon, jun. (*J. Appl. Mech.*, 1959, **26,** Dec., 517–525) Mathematical. The law adopted is $\dot{\epsilon} = (1-E)\dot{\sigma} + \lambda\sigma^n$.

A phenomenological theory for the transient creep of metals at elevated temperatures E. Z. Stowell (*NACA TN*–4396, 1958, Sept., pp.31; from *US Res. Rep.*, 1959, **31,** March 13, 167) An alloy of two metallic phases, each with its own elasticity and viscosity, will exhibit transient creep after application of a constant stress. Four metals show entire families of creep curves to be given by a single set of constants.

Creep-rupture strength of austenitic Cr–Ni–Mo steels in sheet and bar form G. V. Smith, F. Garofalo, R. W. Whitmore, and R. R. Burt (*Trans. ASME, J. Basic Eng.*, 1959, **81D,** Dec., 610–616) It is shown that small variations in C, N, and P in 18%Cr–8%Ni–Mo bar and sheet account for about $\frac{1}{3}$ of the 100 and 1000 h creep-rupture strength in tests at 1100, 1300, and 1500°F.

Correlation of high temperature creep and rupture data H. Conrad (*Trans. ASME J. Basic Eng.*, 1959, **81D,** Dec., 617–628) The problem of correlating creep and rupture data at $T > 0{\cdot}45T_m$ is examined and it is concluded that deformation of metals and alloys is diffusion controlled at these temp.; activation energy is given by that for the self-diffusion of the base metal or alloying constituents, and should be independent of stress. An equation is derived for creep rupture having the form: $\dot{\epsilon}_s = \frac{K}{t_r} \approx A \exp -\left(\frac{\Delta H_D}{RT}\right) \sinh [\sigma/\sigma_0(T)]^n$. Where $\dot{\epsilon}_s$ is minimum creep rate, t_r the time to rupture ΔH_D a diffusion activation energy, σ the stress, and K, A, $\sigma_0(T)$, and n are constants. Data for several superalloys and an Al alloy were found to be in agreement with this equation (55 refs).

The creep of strain gauges as a rheological problem C. Rohrbach and N. Czaika (*Materialprüf.*, 1960, **2,** March, 98–105) The creep of strain gauges and its probable origin is discussed. The creep is attributed to a yielding of the cement as a result of the restoring forces of the strained wire. The authors give a diagram from which the creep of the strain gauge may be deduced. They also discuss the question to what extent strain gauges may be used for measurement of creep.—T.G.

Creep-resisting materials S. Arwidson (*Jernkont. Ann.*, 1959, **143,** (7), 373–413) Creep resisting materials are enumerated, and their properties and applications described. The materials dealt with are: martensitic steels, austenitic steels, and Ni- and Co-base alloys, with a section on cements (11 refs).

A method of treating the results of long-term strength tests V. I. Nikitin (*Zavods. Lab.*, 1959, (12), 1492–1496) The article presents a simpler and more accurate method of determining the breaking load for any given time, based on F. R. Larson and J. Miller's parameter (*Trans. ASME* 74, (5), 1952) $T(C + \log \tau) = p(\delta)$ where T is the absolute temperature; τ, the time to fracture, C constant, and δ the load. In this equation the relation between the product T log τ and temp. at constant load is linear and the tangent of the slope of such a line to the temp. axis is always equal to C. The graph T log τ −T is therefore made up of a family of straight lines, each of which corresponds to a value of the load. This gives the possibility of constructing a convenient diagram for long-term resistance of the material from which it is possible to determine the ultimate long-term strength for any period of time.

Progress reports of investigation of railroad rails R. E. Cramer (*Univ. Illinois Bull.*, 1959, *Reprint series No.*60, April, pp.13) Failures in control-cooled rails and shelly rail studies are included. A summary of 25 failures from 11 causes and of shelly failures in rolling-load tests is presented.

Progress report of investigation of railroad rails and joint bars R. E. Cramer and R. S. Jensen (*Univ. Illinois Bull., Reprint series No.*59, 1958, May, pp.33) Failures in control-cooled rails, rolling-load tests of joint bars, shelly rail studies, and tests on continuous welded rails make up the report.

Some crankshaft failures: investigations, causes and remedies R. Atkinson and P. Jackson (*Shipbuilding Ship Rec.*, 1960, **95,** Feb. 11, 180–181) Failures of Diesel engine shafts are briefly reviewed and the methods of examination are given.

Third International Symposium on hardness measurement in theory and practice W. Bischof and K. Meyer (*Materialprüf.*, 1960, **2,** March, 107–114) The third international symposium on 'hardness measurement in theory and practice' was held at Dortmund on 23–25 Sept. 1959. The present report gives the titles and abstracts of 28 papers by contributors from all over the world. The papers and discussions will be published in full by the VDI, Dusseldorf in the near future.—T.G.

The technique of hardness testing V. L. James (*Sheet Met. Ind.*, 1960, **37,** Feb., 125–135) This final instalment refers to Cu and its alloys.

A simplified calculation of Brinell hardness R. Böklen (*VDIZ*, 1960, **102,** Feb. 1, 128) Brinell hardness H_B is determined from an equation $H_B - H_M = Kx$, where H_M is the Meyer hardness and Kx a corrective term depending solely on the load.

Hardness testing of thin sheet and tinned sheet E. Wondris and V. Seul (*Stahl Eisen*, 1960, **80,** March 17, 356–365) The authors discuss the advantages offered by using the super Rockwell hardness testing for thin sheet and tinned sheet and report on studies of the mechanism involved in the impression of thin sheet, they also discuss the effect of the thickness of the sheet and that of the support on the results. The correlation with other methods and the use of conversion factors are touched upon.—T.G.

Measuring scratch depth (*Ind. Fin.*, 1960, **12,** March, 57) Details are given of an instrument developed by Boeing Aircraft Co. for measuring the depth of scratches ranging from 0·030 to 0·00005 in. The instrument measures the length of the shadow cast by a beam of light at a pre-determined angle.—G.F.

Conversion chart for microhardness values R. A. Slepian (*Mat. Design Eng.*, 1960, **51,** Jan., 123) A graph designed for converting microhardness values to the more commonly used microhardness scales is illustrated.

Hardenability studies on 4340 steel E. R. Klier, V. Weiss, and G. Sachs (*PB* 130766, 1954, Oct., pp.46; from *US Govt. Res. Rep.*, 1959, **31,** June 12, 470) Jominy tests were carried out after various heat treatments. Metallographic constituents were determined and X-ray studies were made.

Some considerations on the problem of brittle fracture O. Werner (*Materialprüfung.*, 1959, **1,** June, 189–200) Modern theories of metal pasticity are discussed, and the contribution of dislocation theory to present knowledge of the causes of brittle fracture is reviewed. Experimental results in the field of internal friction, and the results of electron microscopical investigations are reported, which support theories based on dislocation mechanism, and in which dislocations can be viewed directly (43 refs).

Embrittlement of solid metals in a liquid metal W. A. Morgan (*Met. Treatment*, 1959, **26,** Sept., 333–339) A review of causes of embrittlement and cracking of steels and non-ferrous metals (57 refs).

Micromechanism of brittle fracture in a low-carbon steel G. T. Hahn, W. S. Owen, B. L. Averbach, and M. Cohen (*Weld. J.*, 1959, **38,** Sept., 275s–276s) Tensile data, elastic limit measurements, and metallographic observations of slip, twinning, and micro-cracks, for samples tested in the range room temp. to −250°C, as well as standard Charpy data, are presented for a low-carbon steel in a coarse (ASTM GS No.4) and a fine-grained (ASTM GS No.7) condition.

Brittleness of cobalt-iron alloys C. W. Chen and G. W. Wiener (*J. Appl. Phys.*, 1959, **30,** April, 199s–201s) A mechanism is advanced.

On hydrogen brittleness of steel: III. Metallurgical observations I. Onishi and Y. Kikuta (*Techn. Rep. Osaka Univ.*, 1960, **10,** Jan., 193–208) [In English] The cracks and fracture surfaces formed in embrittled cathodically charged mild steel are described.

Hydrogen embrittlement of plated steel springs J. A. Gurklis, L. D. McGraw, and C. L. Faust (*PB* 151125, 1956, Dec., pp.46; from *US Res. Rep.*, 1959, **31,** April 17, 249) With reasonable care no subsequent heat-treatment or other treatment is necessary after Cd plating.

Brittleness of special steels in the as-cast condition: influence of impurities R. Castro and A. Gueussier (*Rev. Mét.*, 1960, **57,** Feb., 135–148) The influence of structure and composition on the brittleness of an Ni–Cr–Mo steel and an austenitic Ni–Cr steel in the as-cast condition were investigated by means of temp.-impact strength curves. Topics considered were: the effects of mould size, and hence the cooling rate on minor segregation and brittleness; influence of various impurities on intergranular brittleness in heat-treatable Ni–Cr–Mo steels; influence of a homogenizing thermal treatment on brittleness; and the effects of various impurities on the brittleness of as-cast austenitic stainless steels, together with the modifying influence on the results of the presence of 5–10% ferrite in the structure.

Experimental study of solidification cracks in steels: influence of impurities A. Gueussier and R. Castro (*Rev. Mét.*, 1960, **57,** Feb., 117–134) The experimental technique used by the authors is described in detail; one feature is the simultaneous preparation in a single mould of 16 specimens, to reduce the number of tests and the effects of dispersion. Results obtained on Ni–Cr austenitic steels show the influence of primary structure on crack sensitivity; com-

pletely austenitic structures are much more sensitive than mixed austenitic–ferritic structures. The effects of S, B, P, Zr, Nb, Sb, Cu, Ce, Mn, V, Mo, W, Al, Co, Si, and Pb on crack sensitivity are considered. The influence of the results obtained on the welding of stainless steels for elevated temp. use is examined (19 refs).

The mechanism of brittle failure of steel under impact bending of notched specimens B. S. Kasatkin (*Fiz. Met. Metallov.*, 1959, (1), 75–84) The process of brittle failure must be subdivided into two main stages; the first, the pre-failure stage is characterized by plastic deformation and connected with the crushing of grains, twinning, origin, and development of microcracks in the internal microvolumes of the metal. The second stage, corresponding to the actual brittle failure, begins with the formation of plastic cracks in the thin surface layer of the metal arranged at the base of the notch, cracks which are clearly distinguishable from those arising from the deformation of the internal volumes. Thereafter the development of the failure extends by the formation of internal cracks separated from each other by bridges of deformed grains.

Spontaneously induced brittle fractures in welded-steel discs A. Vinckier (*Weld. J.*, 1959, **38**, Sept., 354s–359s) An investigation of the conditions required for spontaneous brittle fracture in welded steel discs is described.

A method of assessing the tendency of steel to brittle failure from the appearance of the fracture E. M. Shevandin (*Zavods. Lab.*, 1959, (12), 1497–1502) The tendency to brittle failure can be assessed not only from impact strength as a function of temp. but also from the appearance of the fracture. On the suggestion made by the author in 1947 one can construct the curves for the work necessary for fracture A and for the relative fibrous fracture B as a function of temp. and find from them the corresponding critical brittle temp. The article explains the main features of the method which can also be extended to assess the safety factor in structural steels from the appearance of the fracture.

Assessment of cold-shortness of structural sheet steel M. I. Kurmanov, I. V. Navrotskii, and Yu. S. Tomenko (*Zavods. Lab.*, 1959, (11), 1370–1372) The threshold of cold-brittleness determined on notched specimens with static loads may be over or under the transition temp. to a brittle state depending on the stress system. Testing is therefore necessary under different stress systems as well as with different stress raisers in order to arrive at comparable figures.

Interstitials and fracture of metals T. E. Scott and A. R. Troiano (*Nature*, 1960, **185**, Feb. 6, 372–373) Hydrogen in steel and delayed brittle failure are considered. The incubation period depends on the time required to accumulate a critical concentration of hydrogen in the region of maximum triaxiality. Notches or other stress raisers bring about these stresses. The crack initiation and propagation processes are discussed.

Friction-testing equipment for small cylindrical specimens G. Rust (*Materialprüfung.*, 1960, **2**, Jan. 20, 17–18) A description of the design and operation of the equipment is followed by a note of the results of the determination of μ for the pair Al–Carbon steel.

Mixed friction as a physico-chemical solid body problem. Influence of metal plasticity on the kinetics of mixed friction A. Knappwost and E. Burkard (*Z. Metallk.*, 1959, **50**, Sept., 540–545) Equations for mixed friction (liquid–solid) in the Coulomb region are derived, and friction experiments with the pairs Fe–Mg, Fe–Sn, and Fe–Al are described. It is shown that avoidance of the Coulomb region by making use of the plastic properties of the material can lead to very low coefficients of friction ($2{\cdot}5 \times 10^{-3}$) in the combination Fe–Al.

Friction and wear of metals to 1 000°C E. P. Kingsbury and E. Rabinowicz (*Trans. ASME(D)*, 1959, **81**, June, 118–122) An investigation is described on the friction and wear of Zn, Cu, Ti, and 1020 steel at temperatures up to 1 000°C. Sliding experiments were carried out using a pin-on-disc friction apparatus incorporated in a metallurgical furnace. At room temp. the results with steel but not with the other metals depended markedly on the atmospheric moisture content. As the temp. was raised to 400°C the friction and wear rate of the steel reached a maximum at 100°C and then diminished while with the other metals the friction remained constant and the wear rate increased. These and other results are discussed in terms of the W/p ratio for the metals, W being the surface energy and p the hardness. It is postulated that materials with low W/p ratios should have favourable friction properties at high temperatures.—D.L.C.P.

The wear of steel rails and tyres in railway service J. Dearden (*Wear*, 1960, **3**, Jan., 43–59) The wear of rails and tyres is reviewed and the economics of the problem as it affects British Railways is discussed. The author also gives numerical values for rail and tyre wear under different conditions and indicates means of reducing wear, for instance by flame-hardening of the tyre soles.—T.G.

A new method for the determination of wear of machine parts M. M. Khruschov (*Wear*, 1960, **3**, Jan., 60–71) The author suggests the use of special marks such as diamond-cut grooves as reference points for assessing the wear of machine parts. Such reference marks enable the duration of wear tests to be cut down and offer the advantage of high accuracy in the estimation of the wear rate.—T.G.

Piston ring friction—rig measurements with low viscosity oils A. de Faro Barros and A. Dyson (*J. Inst. Petroleum*, 1960, **46**, Jan., 1–15, discussion 15–19).

New wear resistant steel is easy to heat treat (*Steel*, 1960, **146**, Jan. 18, 100–101) Properties of a new alloy 'Carpenter Hi Wear 64' are described. The composition is 1·5%C; 2·0%Mn; 0·25%Si, 0·90%Cr, 1·0%Mo; 4·0%W. This alloy has excellent wear resistance at room temp. and is used for punches, dies, shear blades, and for handling abrasive materials.

Nondestructive testing in the steel industry J. B. Austin (*Blast Furn. Steel Plant*, 1960, **48**, Feb., 174–178) A general review of types of test in use, methods of calibration, and the application of non-destructive testing to preventive maintenance. In the system used by United States Steel Corporation, items likely to fail are checked periodically, and frequency of testing is increased in cases where irregularities are observed.

Optical measurement of principal stresses Aeloque (*Engineer*, 1959, **208**, Dec. 4, 754) An account of the clinopolariscope for use with transparent materials.

Optics on machine tools A. Metz (*Maschinenwelt Elek.*, 1959, **14**, Aug., 375–378) Several examples are given of the use of microscopes and telescopic equipment.

Anisotropy of crystals and the structure of the Debye ring in the stressed specimen G. I. Aksenov and V. A. Moschanskii (*Fiz. Met. Metalloved.*, 1959, (6), 847–854) A theoretical analysis of the conditions for the reflection of monochromatic light from the atomic surface of a linearly stressed polycrystalline aggregate, given by one of the authors (Aksenov, *Zh. Fiz.*, 1926, **6**, (2)) has established that elastic deformation must be accompanied by a shift in the Debye line in the X-ray photographs. The present experimental work, by the measurement of the residual stresses, has fully confirmed the theoretical results.

Fluorescent penetrant method for casting inspection A. Lindgren (*Mod. Castings*, 1959, **36**, Oct., 105–110) The application of magnetic ink crack detection methods for the inspection of malleable castings is described.—A.D.H.

Cast epoxy models A. W. Wakeman (*Prod. Eng.*, 1959, **30**, Dec. 21, 60–64) A rapid process for casting models and prototypes is described.

Magnetic and penetrant methods of non-destructive testing H. J. Bezer (*Inspection Eng.*, 1959, **23**, Sept.–Oct., 97–102) An illustrated account of methods.

Investigation of the transformation in alloyed Permalloy M. P. Ravdel' and Ya. P. Selisskii (*Fiz. Met. Metalloved.*, 1959, (6), 885–892) The specimens were of the Fe–Ni alloy nearest to Ni_3Fe (Permalloy) composition, and alloyed with Mo, Cu, and Cr, also with such elements as Mn, Si, W, and V. The electrical resistance and the magnetic properties measured by dilatometer and thermal magnetometer have shown different effects due to the alloy elements, so for example, addition of 3%Mn (on the Fe content), intensifies appreciably the ordering of the Ni_3Fe alloy and this results in a considerable reduction of the electrical resistance. The volume effects also, due to disordering, are more intense than in the binary alloys. The experimental results for the other elements are tabulated.

Magnetic properties of steel as a basis for structure analysis G. S. Tomilov, M. N. Mikheev, and M. F. Pomukhin (*Fiz. Met. Metalloved.*, 1959, (2), 176–181) The relations between the magnetic and the mechanical properties of steel and the microstructure, have first to be investigated before it is possible to design magnetic control apparatus; for example, in low-carbon structural steels with alloy elements, a direct relation between hardness and coercive force is very often observed over the entire range of thermal treatment, and this makes it possible to introduce magnetic control without difficulty. In addition to the existing control methods for the quality of the thermal treatment of steels quenched to martensite and subjected to a high temp. anneal, an automatic magnetic control method is possible for temp. above 220°.

Determination of suitable test conditions for the Magnaflux method G. Beckmann and H. Dähne (*Neue Hütte*, 1960, **5**, Feb., 89–98) The technique and its limits of detection are reviewed, examples are given, and possibilities of developing tests with results acceptable to different parties are discussed.

The effect of finely dispersed precipitates on the coercivity of transformer sheet R. Wagner. (*Acta Met.*, 1959, **7**, July, 523–525) A letter. Tests of 4·1%Si steel with C, Mn, P, S, and Cu in small amounts showed a rise of coercivity at 500°C. Electron micrographs disclose the separation of constituents the sizes of which are estimated and their effects discussed.

Ferrimagnetic resonance line widths and phase distributions in sintered yttrium iron garnets L. G. Van Uitert, F. W. Swanekamp, and S. Preziosi (*J. Amer. Ceram. Soc.*, 1959, **42**, Oct., 471–473).

Domain structure of nickel and iron crystals J. Kocinski (*Acta Phys. Polon.*, 1959, **18**, (3), 169–189).

Surface-integral methods of calculating forces on magnetized iron parts C. J. Carpenter (*Proc. Inst. Elect. Eng.*, 1960, **107**, March, 19–28).

Transformer laminations British Thomson-Houston Co. Ltd (*Iron Steel*, 1960, **33**, Feb., 77–78) An account is given of the manufacture, treatment, and testing of cores produced in cold-reduced, grain-oriented steel.

Exchange anisotropy in an iron–aluminium alloy J. S. Kouvel (*J. Appl. Phys.*, 1959, **30**, April, 313s–314s).

Radiation effects on the anisotropy and magnetostriction of single crystals of several soft magnetic materials including Ni, Fe, Fe_3O_4, and alloys of Ni–Fe, Si–Fe, Al–Fe, Co–Fe and Mo–Ni–Fe R. C. Hall, W. S. Byrnes, and R. G. Crawford (*J. Appl. Phys.*, 1959, **30**, April, 288s–289s).

Anisotropy constants of iron and iron–silicon alloys at room temperature and below C. D. Graham, jun. (*J. Appl. Phys.*, 1959, **30**, April, 317s).

On the nature of defects in the magnetic structure of wüstite W. L. Roth (*J. Appl. Phys.*, 1959, **30**, April, 303s–304s) Neutron diffraction measurements on FeO at 290°K and 4·2°K were made.

Synthesis of a (110) [001] type torque curve in silicon iron C. G. Dunn and J. L. Walter (*J. Appl. Phys.*, 1959, **30**, April, 320s).

Magnetic contributions to the elastic constants of nickel and an Fe-30%Ni alloy at a high magnetic fields G. A. Alers, J. R. Neighbours, and H. Sato (*J. Appl. Phys.*, 1959, **30**, April, 231s–232s).

Nucleation of ferromagnetic domains in iron whiskers R. W. De Blois and C. P. Bean (*J. Appl. Phys.*, 1959, **30**, April, 225s–226s).

Effect of impurities on the temperature dependence of the (110) [001] texture in silicon-iron J. E. May and D. Turnbull (*J. Appl. Phys.*, 1959, **30**, April, 210s–212s).

Influence of plastic deformation on the time decrease of permeability in transformer steel A. K. Smolinski, Z. Kaczkowski, and M. Zbikowski (*J. Appl. Phys.*, 1959, **30,** April, 195s–199s).

Effect of magnetic annealing on the properties of (110) [001] oriented 3¼% silicon–iron strip H. C. Fiedler and R. H. Pry (*J. Appl. Phys.*, 1959, **30,** April, 109s–110s).

Development of metallurgical structures and magnetic properties in iron silicon alloys R. H. Pry (*J. Appl. Phys.*, 1959, **30,** April, 189s–193s).

Temperature dependence of the magnetic properties of nickel–iron alloys J. J. Clark and J. F. Fritz (*J. Appl. Phys.*, 1959, **30,** April, 105s–107s).

Saturation magnetization and Curie points in dilute alloys of iron A. Arrott and J. E. Noakes (*J. Appl. Phys.*, 1959, **30,** April, 97s–98s) Fe–Ti alloys are examined.

Magnetic properties of metals and alloys: Magnetic contribution to the anomalous γ-loop shear of Fe–Al alloys R. Kikuchi and H. Sato *J. Appl. Phys.*, 1959, **30,** April, 92s–94s).

Ferromagnetism of an imperfect crystal by the Ising approximation J. Seiden (*J. Phys. Radium*, 1959, **20,** Nov., 876–889).

Magnetic properties of solid solutions of Fe_3O_4 and $FeAl_2O_4$ S. J. Pickart and A. C. Turnock (*J. Phys. Chem. Solids*, 1959, **10,** July, 242–244) A letter.

Slow domain wall motion in homogeneous vacuum-deposited iron–nickel films R. W. Olmen and E. N. Mitchell (*J. Appl. Phys.*, 1959, **30,** April, 258s–259s).

Anisotropy field measurements of Ni–Fi thin films R. G. Alexander (*J. Appl. Phys.*, 1959, **30,** April, 266s–267s).

The use of polarized neutrons in determining the magnetic scattering by iron and nickel R. Nathans, C. G. Shull, G. Shirane, and A. Andresen (*J. Phys. Chem. Solids*, 1959, **10,** July, 138–146).

Barkhausen effect in nickel–iron films N. C. Ford, jun. and E. W. Pugh (*J. Appl. Phys.*, 1959, **30,** April, 270s–271s).

Influence of substrate processing on the magnetic properties and reproducibility of evaporated nickel–iron films K. H. Behrndt and F. S. Maddocks (*J. Appl. Phys.*, 1959, **30,** April, 276s–277s).

Reduction of saturation magnetization of γ-Fe_2O_3 and Fe_3O_4 by pile irradiation W. E. Henry and E. I. Salkovitz (*J. Appl. Phys.*, 1959, **30,** April, 286s–287s) Studies at 4·2°K are reported.

Magnetic anomaly of magnetite A. S. Mil'ner, E. F. Koftun, and I. N. Popov (*Fiz. Met. Metalloved.*, 1959, (6), 832–836) The magnetic properties of magnetite have been investigated between 75 and 100°K also the effect of a change in the sign of the magnetic anisotropy over the transformation temp. on its magnetic properties at lower temp. When switching on and off a magnetic field or simply cutting it off for short periods in the region of 75°K the magnetization of magnetite previously cooled in a magnetic field falls. This is because the anisotropy constant increases steadily from zero value to 120°K and lower. The value of the jump in magnetization of the magnetite below 100°K, depends on the temp. at which the field was applied during cooling. This effect is due to the non-uniformity of the stoichiometric content of the specimen.

Changes in the Debye temperature during the ordering of Fe_3Al alloy S. A. Nemnonov and L. D. Finkel'shtein (*Fiz. Met. Metalloved.*, 1959, (6), 944–945) Since the Debye temperature of pure metals in the annealed and deformed conditions is the same, it may be concluded that, by itself, deformation, not connected to redistribution of the alloy element, does not result in a change in the Debye temperature, and that such a change in the Fe_3Al alloy may possibly be ascribed to the effect of disordering.

A comparison of static and microwave measurements of magnetocrystalline anisotropy in cobalt manganese ferrite R. F. Pearson and R. W. Teale (*Proc. Phys. Soc.*, 1960, **75,** Feb. 1, 314–316).

Magnetic properties of hematite single crystals. I: Magnetization isotherms, antiferromagnetic susceptibility and weak ferromagnetism of a natural crystal S. T. Lin (*Phys. Rev.*, 1959, **116,** Dec. 15, 1447–1452) Measurements from 488°K to liquid He temp. are made.

Measuring magnetic induction in electrotechnical steel with a magnetic field strength of 10 amp turns/cm G. L. Gornshtein and M. N. Leshchiner (*Zavods. Lab.*, 1959, (12), 1427–1428) Rectangular 500×30 mm specimens in batches of 10 kg are tested. Three special permeameters of special construction are described and readings obtained on all three are compared.

Quantitative determination of texture in cold-rolled transformer steel V. V. Druzhinin and I. P. Kudryavtsev (*Zavods. Lab.*, 1959, (12), 1425–1427) Magnetic induction was measured on 45×160 mm sheets at 25 amp. turns/cm. Then strips with rather larger grains were measured for magnetic induction at B_{25} from 20000 to 17000 Gauss. The specimens were then etched electrolytically before the crystal orientation was optically determined. It is possible to conclude from the findings that even for specimens with low induction there are no crystals with disordered orientation. For a quantitative determination of texture a relation must be established between the amplitude of harmonic curves and the dispersion angles.

Efficient methods of assessing the magnetic properties of cold-rolled transformer steel V. G. Borisenko (*Zavods. Lab.*, 1959, (12), 1422–1425) The possibility of using the value of the coercive force to classify sheets of cold-rolled transformer steel according to their magnetic properties was undertaken experimentally. By measuring the coercive force on whole sheets and then cutting the sheets into standard specimens and testing in 10 kg batches on the Epshtein differential apparatus it has been shown that for 0·35 mm thick sheets, there is a sufficiently close relation between coercive force and total specific losses within an induction range from 10000 to 17000 Gauss and that this can be used to sort out low grade sheets.

Non-destructive methods of investigating electro-technical sheet steel R. I. Yanus (*Zavods. Lab.*, 2959, (12), 1421–1422) Modern techniques make possible the construction of apparatus to measure any magnetic property on whole sheets with the required precision. But the greater the precision, the more complicated and costlier the apparatus. These considerations are the subject of this article by the author from the Institute of Metal Physics of the Academy of Sciences USSR.

Determining the properties of electrotechnical steel (Discussion summary) (*Zavods. Lab.*, 1959, (12), 1415–1417) In spite of the introduction of new control methods, wattmeter and ballistic methods remain and will probably remain for a long time, the main methods for these steels. Nevertheless a standard wattmeter or ballistic method of determining the properties does not yet exist. Sample taking according to GOST 802–58 is not always satisfactory. In view of the great expansion in the use of electro-technical steels (especially thin-rolled) new electrical and magnetic characteristics are needed. In elaborating the technology of the production of cold-rolled textured steels, in addition to the electrical and magnetic characteristics, new metallographic control methods are needed for determining the crystallographic and magnetic anisotropy.

Magnetic metallographic analysis of R18 high speed steel E. M. Strunina and M. N. Petrova (*Zavods. Lab.*, 1959, (9), 1092) The method is based on the fact that colloidal magnetic particles applied in the form of a paste on a microsection concentrate themselves on ferromagnetic phases so that the nonmagnetic particles are set free. A pattern then appears on the microsection which makes it possible to study under the microscope the distribution of the ferromagnetic and nonmagnetic phases.

Magnetic domains in cementite M. Hillert and N. Lange (*Jernkont. Ann.*, 1959, **143,** (7), 414–425) The zebra-like pattern observed in cementite when observed under polarized light with crossed polars is discussed, and explained on the assumption that the pattern represents the magnetic domain structure of cementite. Experiments to verify this assumption are described; thin specimens were mounted in the gap of an electro magnetic, and the effect of changes in the magnetic field in the pattern was observed under the microscope.

Internal ferromagnetic resonance in magnetite J. C. Anderson and B. Donovan (*Proc. Phys. Soc.*, 1960, **75,** Jan., 149–151).

Rotational hysteresis loss in grain-oriented silicon–iron W. F. Archenhold, H. F. Sandham, and J. E. Thompson (*Brit. J. Appl. Phys.*, 1960, **11,** Jan., 46–49) An improved torque magnetometer was used.

Rimmed steel magnetic sheets—how they were produced for the proton synchrotron G. Odone and B. Sommacal (*J. Met.*, 1960, **2,** Feb., 156–157) An outline of the manufacture and magnetic properties of steel sheet used for the construction of magnetic blocks in the 25-bev proton synchrotron of CERN at Geneva.

Magnetic anisotropy of Fe^{++} ion in siderite A. Mookherji and S. C. Mathur (*Indian J. Phys.*, 1959, **33,** Sept., 410).

Defectoscopic investigations of rails in magnetic fields in motion. 14. The spectra of signals from certain defects V. V. Vlasov and G. G. Ushakova (*Fiz. Met., Metalloved.* 1959, (6), 837–841) The frequency of the corresponding maxima of spectral densities of signals from defects, with the same rate of motion of the defectoscope, are practically the same. The spectra of signals, with increasing velocity of the defectoscope change and the maxima of spectral densities are displaced towards higher frequencies.

Defectoscopic investigation of rails in magnetic fields in motion. 16. Studying the appearance of the facets by their reaction in the primary field V. V. Vlasov and L. I. Korobkova (*Fiz. Met. Metalloved.*, 1959, (6), 937–939) The experiments have shown that the defects corresponding to the transverse crack in the head of the rail in the zone of the relatively weak external fields shows up better with a coil placed on the second, S pole of the electromagnet than on the first, the N pole.

Defectoscopic investigations of rails in magnetic fields in motion. 17. The study of current in the rail when excited by the transverse field V. V. Vlasov (*Fiz. Met. Metalloved.*, 1960, (1), 25–30) A generator is described for inducing eddy currents in rails by means of a transverse magnetic field. The immediate changes in the magnetic field without motion have been studied. An approximate calculation is given for the value of the eddy currents induced in the rails during the motion of the generator over them.

Solubility of carbon in iron as determined by the magnetic aftereffect J. Singer and E. S. Anolick (*J. Appl. Phys.*, 1959, **30,** April, 193s–194s) A slightly lower value was found than that derived from internal friction measurements: (0·006% at 600°).

The electrical properties of Armco iron and beryllium F. K. Lampson (*NEPA*–1860; *PN* 140790, 1951, May, pp.9; from *US Res. Rep.*, 1959, **32,** Aug. 14, 238) Measurements up to 2450°F were made on Armco iron and on Be from 2000°F.

The Lorenz constant for steel P. G. Morgan (*Iron Steel*, 1960, **33,** Jan., 28–30) A review is given of recent Russian work on the effects of temp. composition and structure on the Lorenz constant for steels and two high-temperature alloys.

An introduction to ultrasonic testing. IV W. E. Schall (*Inspection Eng.*, 1959, **23,** Sept.–Oct., 103–106) The wall thickness gauge, the flaw locating rule, and the interferometer and their uses are outlined.

Characteristics of the ultrasonic radiation field, with special reference to immersion testing J. Krautkrämer (*SS News*, 1960, March, 4–11) The properties of ultrasound and flaw-size and the immersion scanning technique are reviewed.

Ultrasonic testing of welds D. O. Sproule (*Brit. Weld. J.*, 1959, **6,** Oct., 470–479) Ultrasonic procedures are reviewed. The measurement of echo-intensity and surface conditions are considered. Main principles such as the reflection at boundaries in solids, pulse length,

frequency, and resolving power are discussed with illustrations.—U.E.

Ultrasonic testing of welds R. Hornung (*Sulzer Techn. Rev.*, 1959, **41**, (1), 81–86) A system for rapid testing in the workshop and in the field is described. It detects even small flaws and makes pressure testing of pipe lines unnecessary.

Ultrasonic testing and sorting of welded tubing M. H. Allard (*Autom. Prog.*, 1959, **4**, Dec., 396–399, 401) Principles of ultrasonic testing are outlined, and a method of 100% ultrasonic testing of welded tube, operating at the same speed as the production line, is described. The equipment is synchronized with the saw, and marks the position of flaws.

Investigation on a cast steel ship's propeller for the interpretation of defects shown by ultrasonic testing K. Fellcht, A. Walzel, and H.-J. Drese (*Neue Hütte*, 1960, **5**, Feb., 99–105) A detailed examination of the propeller, previously shown by ultrasonic testing to be defective, showed that micro-cavities in two positions had been accurately delineated by the instrument, but other faults, such as blowholes and cracks, had not been detected. It was concluded that the faults detected, occurring at the transition from blade to hub, were normal and would not require scrapping the propeller, while the more significant faults had not been detected.

The railway axle ultrasonic test assembly J. Krautkrämer and H. Krautkrämer (*SS News*, 1960, March, 12–14). **Ultrasonic rail test assembly** (15–16).

A nondestructive method to detect pipes and cavities in hot steel blooms during the rolling-process by means of betatron, X-ray-image-intensifier and television-setup W. Lueckerath, K. Fink, R. Flossmann (*Nondest. Testing*, 1960, **18**, Jan.–Feb., 27–34) The range of 4–8 in. steel is covered. Pipes with a width of $\frac{1}{8}$ in. in the direction of the beam are made visible. Slag-free and slag-containing shrinkage cavities are not distinguished. The max. speed is 40 ft/min.

Radiographic inspection techniques for weld examination R. Halmshaw (*Brit. Weld. J.*, 1959, **6**, Oct., 456–466) The detectability of defects with the radiographic technique is discussed. A method for calculating the dimensions of just-detectable cracks is given.—U.E.

Simple electronic scanning system for the X-ray microanalyser A. Bernard, D. Bryson-Haynes, and T. Mulvey (*J. Sci. Instr.*, 1959, **36**, Oct., 438–444) An electrostatic deflector system for manual or electronic control of an electron probe is described with an account of its use for examining the spark-hardened layer on a steel pip. Statistical graininess limitations of the image are noted.

Notes on the use of radiographic inspection in naval construction F. Sanchez Laulhé and C. Garcia Mauriño (*Cien. Tecn. Solda*, 1959, **9**, May–June, 1–10) Some data are given on the radiographic inspection services at the shipyards of Empresa Nacional Elcano of Seville.

X-Ray investigation of austenite during intermediate transformation L. S. Palatnik and I. A. Tananko (*Fiz. Met. Metalloved.*, 1959, (6), 842–846) A method is explained permitting thermal treatment directly in the X-ray room and thus recording the stages of the $\gamma \rightarrow \alpha$ transformation in the intermediate domains. From the analysis of the different interference spots (311) of the γ-phase it has been shown that the thermal transformation of austenite in the intermediate domains results in a change of structure which is due to a refining of the crystals, i.e. to much finer crystal elements (blocks, fragments) and to their disorientation with the appearance of microstresses.

X-Ray analysis of small amounts of residual austenite G. M. Rovenskii and M. I. Ermolova (*Zavods. Lab.*, 1959, (11), 1330–1332) With low austenite contents it is not possible to measure the intensity of radiations very weak as compared with those from the martensite. It is however possible to weaken the intensity of the martensite radiation by a local absorber so that the intensity of the austenite radiations could be measured. This is the principle of the proposed method.

The change of lattice constants of iron-rich iron–aluminium alloys with heat-treatment F. Lihl and E. Burger (*Arch. Eisenh.*, 1960, **31**, Feb., 129–132) Fe–Al alloys containing 8·6–30·4%Al were studied by X-ray analysis after various heat-treatments. It was found that the discontinuities in the lattice-constant concentration curve found by Bradley and Jay are not real, but are caused by variation in the cooling conditions. Alloys with more than 12%Al show such a tendency to order that it was not possible to obtain them in the disordered state.

The theory and practice of electronic testing of winding ropes T. Harvey and H. W. Kruger (*Trans. S. Afr. Inst. Elect. Eng.*, 1959, **50**, June, 126–166, discussion 166–181) A very complete account of work for the S. African mining industry.

X-Ray and neutron scattering from electrons in a crystalline field and the determination of outer electron configurations in iron and nickel R. J. Weiss and A. J. Freeman (*J. Phys. Chem. Solids*, 1959, **10**, July, 147–161).

X-Ray diffraction pattern of carbide in low-carbon iron–silicon alloys G. Konoval, L. Zwell, L. A. Gorman, and W. C. Leslie (*Nature*, 1959, **184**, Dec. 12, 1862–1863) A needle-like constituent of 3·25%Si iron metastable to cementite was examined. Raising the C to 0·06%, quenching and ageing gave a structure which produced an X-ray diffraction structure which was not, however, identifiable.

Measurement of tinplate thickness using fluorescent X-rays excited by a radioactive source J. F. Cameron and J. R. Rhodes (*Brit. J. Appl. Phys.*, 1960, **11**, Jan., 49–52) Tritium 'bremsstrahlung' was used as source.

Radio-isotopes and Soviet steel A. B. Tesmen (*Met. Prog.*, 1959, **76**, Sept., 115–120) The use of radioactive tracers in blast-furnace burden investigations, control of blast-furnace refractory wear, automatic control of charge level, and steelmaking investigations in the Soviet Union are described.

Radiation damage in metals D. Peckner (*Mat. Design Eng.*, 1960, **51**, Jan., 83–93) The causes and effects of radiation damage on various metals are explained and a method of restoring the properties of damaged specimens by post-irradiation heat treatment is suggested.

Radiation damage: its mechanisms and manifestations in reactor materials R. J. Bosnak (*J. Amer. Soc. Nav. Eng.*, 1960, **72**, Feb., 15–23) A general account of the theory of radial tion damage and its effects on the mechanica-properties of substances including steels.

The technical applications of radioisotopes G. Pröpstl (*VDIZ*, 1960, **102**, Feb. 1, 138–141) A review of the literature dealing with applications to textile engineering, chemical engineering, metallurgy, materials testing, and changes of properties of materials resulting from irradiation (96 refs).

Comparison of parameter methods for extrapolating high-temperature data R. M. Goldhoff (*Trans. ASME J. Basic Eng.*, 1959, **81**, Dec., 629–644) Three parameter methods are compared statistically for their ability to reproduce the isothermal data on which they are based, and comparisons are made with long-time extrapolations based on direct plotting of experimental data. In general, predictions of long-time working results will give better results than long extrapolations on double-logarithmic plots.

Cast and wrought alloy steels. Elevated temperature properties R. K. Buhr and W. A. Morgan (*Mod. Castings*, 1959, **36**, Oct., 61–67) The mechanical properties of four steels containing 0·38%C, 0–1·68%Ni, 1–5%Cr, 0·4–1·7%Mo, and 0–0·32%V were determined in the cast and wrought conditions over the temp. range 70–1000°F. Although at room temp. the ductility of the cast samples was low it was comparable with that of the wrought condition at elevated temp. The steels show secondary hardening which is due to precipitation of alloy carbides.—A.D.H.

Effect of prior cold work on the high temperature properties of a chromium molybdenum steel P. Shahinian (*PB* 139594; 1955, June, pp.21; from *US Res. Rep.*, 1959, **31**, June 12, 470) Measurements at 370–540°C after 0, 8, 15, and 39% reduction of cross-section are described.

Method for determining the heat resistance of cast iron N. N. Aleksandrov (*Zavods. Lab.*, 1959, (11), 1335–1336) The only method of obtaining comparative results for the heat resistance of cast irons with any form of graphite inclusion is by assessing their resistance to scale formation from the increase in weight. To determine the other characteristics of heat resistance, i.e. the tendency for growth, it is best to use cylindrical specimens with pins of 1Kh18N9T steel screwed into the ends for measuring on an extensometer.

The determination of the hot yield point of steel F. Körber and A. Pomp (*SLA* 59–20476; translated from *Trans. Kaiser–Wilhelm–Institut Eisenforschung Mitt.*, 1930, **12**, (12), 165–169; from *US Techn. Trans.*, 1960, **3**, Jan. 8, 51) Conditions of loading are studied.

Assessment of the heat resistant properties with time of heat-treatment resistant steels A. D. Stanyukovich and N. D. Zaitsev (*Zavods. Lab.*, 1959, (9), 1101–1106) This assessment is based on testing, at a constant rate of deformation. Plastic properties change with temp. and are plotted on semilogarithmic coordinates showing a V-shaped ourve. A reduction in the rate of deformation shifts the curve towards lower temp. From the V-shaped family of curves it is possible to find the minimum deformation capacity of the steel in relation to temp. The relation between the relative elongation and the rate of deformation at constant temp. can be expressed as a power function. The deformation capacity of heat-resistant steels may be approximately expressed by a plasticity diagram constructed by extrapolating the experimental data.

The effects of solute elements on the strength properties of iron- and cobalt-base alloys A. K. Wolff, E. E. Underwood, R. E. Maringer, and G. K. Manning (*WADC Techn. Rep.*, 59–258; *ASTIA Doc. AD* 215432, *PB* 161043, 1959, April, pp.63) Fe–Cr alloys were investigated for high-temp. deformations in regard to temp., strength, stability criteria, lattice strain, and valency changes.

Investigating the properties of titanium alloys IV. Properties of the ternary alloy of the Ti–Al–Fe system I. I. Kornilov, E. N. Pylaeva, and M. A. Volkova (*Fiz. Met. Metalloved.*, 1959, (2), 182–186) The heat resistance of Ti alloys is a function of the chemical composition and of the structure of the ternary Ti–Al–Fe system. The highest resistance is shown by alloys in the region of saturated or super-saturated Ti ternary α-solid solutions. These solutions contain from 3 to 8% Al and 0·5%Fe.

Design and construction of an adiabatic high-temperature calorimeter E. Scheil and W. Normann (*Z. Metallk.*, 1960, **51**, March, 159–164) The calorimeter is intended for measuring the heat of mixing of metals at the temp. of mixing. The reasons for choosing the adiabatic modification are explained. The design and construction is then described, the crucible proper is made of graphite surrounded by the calorimeter body of silver. Unfortunately, trouble with the insulation of the various electrical parts limited the application of the instrument to temp. below 450°C, i.e. to such low-temp. alloys as Pb–Bi.—T.G.

Measurements of thermal properties I. B. Fieldhouse, J. C. Hedge, and J. I. Lang (*WADC–TR–58–274, AD–206892*, 1958, May, pp.87; *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1214) Stainless steels 316, 347, and non-ferrous materials were examined. Conductivity, sp. heat, and linear expansion were measured from 100–3000°F or to the m.p.

The true specific heat of iron, nickel and chromium at high temperatures: New methods for the determination of the true specific heat of metals at high temperatures H. Lange and R. Kohlhaas (*Forsch. Nordrhein–Westfalen*, 1960, (797), pp.115) This is a review of work carried out by various authors at the University of Cologne on the determination of the true specific heat of metals at high temp. The original results, reported elsewhere, are summarized and discussed (83 refs).—T.G.

Apparatus for measuring thermal conductivity of ceramic and metallic materials to 1 200°C W. H. Sutton (*J. Am. Ceram. Soc.*, 1960, **43**, Feb. 1, 81–86) Steady-state apparatus for measuring thermal conductivity by water calorimetry is described, with accuracy

of approx. 1–3% in range 30°–300°C and 3–10% in range 300°–1200°C. Curves for samples of alumina, stainless steel, and graphite are given.—S.H.–S.

Modern topics of diffusion in metals N. Kunitomi (*Met. Phys.*, 1959, **5**, Nov., 238–252) [In Japanese] A review with captions in English.

Directional deforming hydrogen diffusion in the solid solution and the strength of metals K. V. Popov and V. A. Yagunova (*Fiz. Met. Metalloved.*, 1959, (2), 187–192) Various hypotheses are current regarding H_2 embrittlement. The present work aims at assessing the H_2 brittleness due to internal pressures in the microcavities because of directional hydrogen diffusion during plastic deformation. Present investigations point to the fact the diffusion rate is greater in deformed metal and that the time of deformation often exceeds 1 min. It is therefore reasonable to assume that the pressures in the original foci of the incipient cracks attain considerably larger values than may be calculated from the well-developed cracks and that the diffusion of H_2 into the micro-defects is one of the causes of steel brittleness.

Solubility and the displacement of carbon in ferrite L. S. Lifshits (*Fiz. Met. Metalloved.*, 1959, (1), 31–37) Displacement of carbon atoms in ferrite and all other diffusion processes take place when there is a difference in concentration in the ferrite. The difference in the carbon concentration in the various portions of the specimen is conditioned by the non-uniformity of the carbides of the alloy elements. With a high carbon content in a steel containing carbide-forming elements, the equilibrium conditions between combined and free carbon cause an increase in the carbon dissolved in the ferrite.

Nitrogen in high-manganese steels L. I. Levi and A.A.Zhukov (*Lit.Proizv.*, 1959, (12), 24–27) The contradictory nature of the effect of N in iron alloys is explained by the different nature of the solutions formed by it. In liquid iron and in ferrite the solubility increases with increasing temp. which is typical of solutions of the Fe–C and Fe–H type. In case of austenite, the reverse behaviour is observed. N_2 exists in austenite in the form of crystalline compounds which dissociate with increasing temp. Mn, like γ-iron, has a negative solubility coefficient for N_2 forming with it solutions with chemical type bonds. Furthermore it stabilizes the austenite in steels and diminishes the intermittent reduction of solubility of N_2 during the solidification of the melt.

Effect of palladium additions to iron, on its permeability by hydrogen V. I. Arkharov and A. A. Kralina (*Fiz. Met. Metalloved.*, 1959, (1), 45–52) The H_2 permeability of iron is increased by additions of 0·5%Pd and this permeability occurs mainly at the boundaries. Permeability into the volumes of the crystals is conditioned mostly by the joints between blocks in the sub-crystal structure. The effect of Pd additions may be explained by its adsorption.

The partitioning of alloying elements in malleable irons G. Sandoz (*NRL Report* 5268; *PB* 151413, 1959, Feb., pp.26) Partitioning of several alloying elements between the cementite and austenite phases of irons during first-stage graphitization has been determined. V, Cr, Mo, and Mn (not as MnS) concentrate in the cementite phase, and small amounts of Si, Cu, Ni, Co, and Al are found. The mechanism by which alloying elements influence graphitization is discussed (33 refs).

Effect of titanium on the properties of boron-containing steel Ya. E. Gol'dshtein and L. L. Pyatakova (*Lit. Proizv.*, 1959, (12), 28–31) Different types of boron steel were investigated, after being remelted in industrial acid electric furnaces and in basic OH furnaces. The effect of different melting conditions, of the nature of the deoxidizing agents, and of the residual Al on the hardenability in depth are tabulated. Similarly findings for different deoxidizers and micro-alloying on impact strength, on mechanical properties and cold shortness are shown. The combined effect of Ti and B on grain growth was also investigated. As a conclusion, it is pointed out that the advisability of Ti additions must be decided having regard to the intended purpose of the steel and to the conditions of its subsequent mechanical and thermal treatments.

Titanium-bearing steels for nitriding L. Gillemot and T. Tömöry (*Gép*, 1958, **10**, (5–6), 177–184; from *Hungarian Techn. Abs.*, 1959, **11**, (3), 92) Ti increases rate of nitriding so that the process takes 5 h instead of 50 h. Advantages for parts subject to alternating stresses are specified.

Precipitation hardenable iron aluminium alloys W. J. Buehler and C. G. Dalrymple (*NAVORD Rept.* 6252; *PB* 140383, 1959, Jan., pp.25; from *US Res. Rep.*, 1959, **32**, 238) Hardening effects of Ti, Cu, and Ni on an alloy with 8%Al were measured.

Special cast irons P. Lemal (*Rev. Univ. Mines*, 1960, **16**, Feb., 81–85) The author reviews a number of special irons and their uses in chemical industry. The effects of the main alloying additions on high temp. behaviour, and resistance to chemical attack and abrasion are outlined.

Study of some properties of austenite cast irons containing between 5 and 10% nickel A. Meallier (*Fonderie*, 1959, (158), March, 107–124) The properties are examined of austenitic cast irons containing 5–10%Ni and 4·4–6·9%Mn, the 5%Ni alloy also containing 3·6%Cu.

Quality requirements of super-duty steels J. F. Elliott (*AIMME, Elect. Furn. Steel Conf. Proc.*, 1958, **16**, 47–53) The author summarizes a number of papers presented at an AIMME Conference on super-duty steels, outlining the environments in which these steels are used, and the properties and structures required to withstand those environments.—G.F.

Quality requirements of super-duty steels B. M. Shields (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 53–65) The author reviews the part of the AIMME Conference on super-duty steels which deals with the problem involved in their production. He considers first air-melting in conventional arc furnaces, and then discusses the advantages and disadvantages of the various new vacuum techniques.—G.F.

The selection of carbon steel bars W. de Lima e Silva (*Eng. Min. e Met.*, 1957, **25**, Jan., 31–34; March, 125–126) [In Portuguese] The author discusses the classification of carbon steels with reference to their method of manufacture, composition, quality, microstructure, and use. Each of these factors is examined briefly.—P.S.

Carbon steels for tools—steels for tools and also for construction F. J. Mariñas (*Met. y Elect.*, 1960, **24**, Jan., 200–205) A survey of their varieties, composition, properties; and structure classified testing.—S.H.–S.

Special low-temperature steels P. Zehringer (*Rev. Univ. Mines*, 1960, **16**, Feb., 77–80) An outline of the problems and nature of brittle fracture is followed by an account of the behaviour of plain C, 5 and 9%Ni, austenitic, Cr–N, Ni–Cr–N, Mn–N, and Cr–Mn–N steels at low temp.

Engineering alloy steels: grades and general technical requirements (*GOST* 4543–57, from *US Techn. Transl.*, 1959, **2**, Nov. 6, 668).

Nickel steels F. Javier Mariñas (*Met. Elect.*, 1956, **20**, June, 102–108; July, 60–69) The effect of Ni upon the mechanical properties, microstructure, and eutectoid composition of steel is described. Heat-treatment is discussed and typical compositions, applications, and uses are given.—P.S.

The rational classification of the quality constructional steels R. Calvo Rodes (*Inst. Hierro Acero*, 1959, **12**, Oct.–Dec., 277–287) The author's aim is to classify these steels into a group such that each member has accurately defined applications and thus avoid unnecessary multiplicity. Deformation, fracture, and mechanical behaviour of metals are discussed; steels as high strength materials are considered. The high strength steels for this group fall in the range, 80–140 kg/mm² tensile strength and must cover sizes of up to 6 in. dia. Fifteen steels, with three grades of strength are thus needed. Factors involved in the choice of such steels are discussed; the three key ones are hardenability, M_s temperature, and potential temper hardness.—P.S.

Wear-resistant cast steels Viallon (*Usine Nouv.*, 1959, **53**, Nov. 31, 29–31) From the stand-point of mechanical wear, the author examines the factors which should determine the choice of a steel; he describes the principal classes of cast steels; studies some concrete examples, and concludes with some advice to the ultimate users of specially wear-resistant steels with remarks on turning, planing, and drilling of Si–Mn and Mn steels.—S.H.–S.

Practical estimation of composition balance and ferrite content in stainless steels L. Pryce and K. W. Andrews (*JISI*, 1960, **195**, Aug., 415–417) [This issue].

New high-strength stainless steels: Jethete series is based on 12%Cr United Steel Co. Ltd (*Brit. Eng. Transport*, 1960, **42**, March, 269–270; *Met. Treatment*, 1960, **27**, Jan., 2).

Stainless steels for service at elevated temperatures (*Brit. Steel.*, 1960, **26**, Jan., 18–19) A description of the properties and application of the Jethete series developed by United Steel Companies Ltd.

Special steels in the chemical industry R. Berger (*Rev. Univ. Mines*, 1960, **16**, Feb., 55–76) The author examines the possible structural reactions in Fe–Cr and Fe–Cr–Ni alloys, and discusses the effects of minor constituents. The resistance to corrosion of Fe–Cr and 18/8 alloys is then studied, followed by a discussion of the chemical stability of Fe–Cr–Ni–Mo, with particular reference to the action of dilute H_2SO_4 and boiling HNO_3.

New ultra high strength steel United States Steel Corp. (*Mat. Design Eng.*, 1960, **51**, Feb., 9, 11) An account of 'USS Strux' and its mechanical properties.

Tough steel features high-strength-to-weight ratio (*Steel*, 1960, **146**, Jan. 18, 96) The properties and use of 'USS Strux' high strength alloy steel are outlined. The steel (0·4 to 0·5%C) can be forged, machined, and heat treated to develop strengths of 280000 to 300000 psi. It is used in the landing gear of the A3J Vigilante, a new jet plane for aircraft carrier service.—D.L.C.F.

Fill many ferrous needs with two alloy steels (*Iron Age*, 1959, **184**, Oct. 8, 110–112) It is suggested that many factories could operate with only two alloy steels one through-hardening (AISI 4340) and one for case-hardening (AISI 4615–20). The special properties of these steels and the reasons for their selection are discussed.—A.G.

Cr–Mo–V low-alloy steels for service at elevated temperatures (*Alloy Met. Rev.*, 1959, **8**, Dec., 2–6) A review of German and US grades.

METALLOGRAPHY

Method of metallographic analysis of the transition-type stainless steels A. S. Tarantova, G. G. Solov'eva, and L. M. Pevzner (*Zavods. Lab.*, 1959, (9), 1089–1091) The steels investigated are the ageing stainless steels of the austenite–martensite transition class (EI 904, EI 925, etc.). After austenization at temp. corresponding to the complete dissolution of the chromium carbides and then cooling in air, the structure of the transition steels is mainly austenitic, sometimes with small amounts of martensite. The authors describe the best method for preparing microsections etching, polishing, etc.

The thermodynamic basis of equilibrium diagrams F. Munoz del Corral (*Inst. Hierro Acero*, 1956, **9**, April–June, 686–700; July–Sept., 1014–1020; Oct.–Dec., 1120–1128; 1957, **10**, Jan.–March, 67–71; April–June, 187–194; July–Sept., 300–319; Oct.–Dec., 412–428) [In Spanish] The author deals with fundamental principles and discusses phases, their co-existence, mixtures of ideal or real gases, equilibria in various binary systems, and concludes with a brief discussion of ternary systems.—P.S.

Methods of determining grain size F. Schückher and L. Janerot (*Jernkont. Ann.*, 1959, (9), 593–614) Various methods are described and compared. Methods used for production purposes in Sweden are the ASTM method based on the arithmetical mean grain area and the JKM method which depends on the determination of the grain size group which covers the greater part of the visible area. In the case of homogeneous structures

without grain size anomalies, it is shown that grain size distribution in two and in three dimensions can be approximately described as a logarithmic normal distribution, and the grain size structure can be described in terms of geometric mean and scatter (26 refs).

The photodensitometer for the structural analysis of metals and its application to cast iron T. Kusakawa, K. Nakamura, and S. Iijima (*Imono*, 1959, **31**, Sept., 800–805) Relative areas of graphite and ferrite are measured from micrographs and found to correlate with composition and mechanical properties.

The microscope in engineering A. E. Bingham (*J. Jun. Inst. Eng.*, 1959, **69**, Aug., 359–382) A general account of apparatus with some account of uses, generally metallographic.

Optical microscopy: a key to metallurgical progress E. E. Thomas (*Bell Lab., Record*. 1959, **37**, Dec., 457–463) The phase concept, and their recognition under the microscope is considered. The examples are non-ferrous.

Apparatus for tilting extraction replicas through large angles in the electron microscope G. R. Booker and J. Norbury (*J. Sci. Instr.*, 1959, **36**, Aug., 368–369).

Rotating sample holder for microscopic observation of whiskers P. J. Bryant and P. L. Gutshall (*Rev. Sci. Inst.*, 1959, **30**, Dec., 1126–1127).

A carbon arc image furnace and microscope for studies at high temperatures H. W. Newkirk and B. B. Brenden (*J. Amer. Ceram. Soc.*, 1960, **43**, March, 146–150) A description of a 10 kW furnace with 12 in. mirror is given.

Physical investigations at high temperatures of fuel ashes, slags and ceramics with the hot-stage microscope A. Metz (*Radex Runds.*, 1959, (4), 612–616) The instrument is illustrated, its design and operation described, and some examples are given of its use.

Investigating the surface of fractures with the electron microscope A. G. Orlov and L. M. Utevskii (*Zavods. Lab.*, 1959, (9), 1084–1087) A description of the procedure, etching, preparation of the surface, and replicas, etc.

Diamond abrasive eases metallographer's task E. C. Olden (*Steel*, 1960, **146**, Feb. 8, 80–81) A method for preparing metal samples for examination under the microscope which retains flatness of the surface at edges and softer areas, and does not smear soft inclusions is described. Key to the method is the use of two diamond pastes, the first with 5 to 8 μ, the second with 0–1 μ particles.—D.L.C.P.

Study of the depth to which deformation proceeds during the preparation of metallographic sections F. Lilh and H. Mayer (*Z. Metallk.*, 1960, **51**, March, 186–190) The aim of the study was to determine quantitatively the depth at which a deformation resulting from the preparation of metallographic sections can be traced. The tests were performed on copper and a mild steel, and show that samples ground wet take a deeper impression than those ground dry. Polishing with Al_2O_3 leaves a rather deeply deformed surface layer. Electrolytic polishing removes the deformed layer completely. The mechanism involved in the deformation is different for copper and steel.

The etching of double and treble phosphide eutectics in cast iron Ya. N. Malinochka and N. G. Osada (*Zavods. Lab.*, 1959, (12), 1523–1525) In etching with an alkaline solution of potassium ferricyanide (10 g of $K_3Fe(CN)_6$ and 4 g NaOH in 100 g water) the dark cast iron takes on the colour in a boiling solution before the phosphides and cementite. If etched in the cold or in a solution heated to 40–60° only the phosphides are coloured light orange to dark brown depending on the temp. of the solution and the duration of etching.

New method of study of the gamma grain of steels and the role of aluminium nitride J. Pomey and G. Vigneron (*Génie Civil*, 1959, **136**, D3c. 15, 513–514) The surface of a steel sample is mirror-polished and heated to the desired temp., then a metal or alloy is fused to the surface which forms mixed crystals with it while cooling, and the preparation is examined microscopically.—S.H.–S.

Method of revealing the dendritic structure of carbon steel I. E. Levshin (*Zavods. Lab.*, 1959, (11), 1333–1334) The method is based on combined etching. The specimen, first cleaned with a coarse and a smooth file is ground with emery paper 280 or 320 and degreased in alkali or alcohol, ground again with M14 or M10 paper then polished with a chromium oxide-impregnated cloth. Etching is carried out at 18–20° in a 10% aqueous copper Heyn solution ($CuCl_2 2NA_4Cl$) and then in an aqueous solution of picric acid. During the process the surface of the specimen is rubbed every 30–50 sec. with a cotton-wool pad. The copper deposit is removed if necessary with ammonia. The specimen is again etched in a solution of picric acid for 5–6 h but has to be rubbed with cotton wool every 15–20 min. The specimen is again immersed in the Heyn solution for 30–50 sec. and finally washed with hot water and dried.

A metallographic technique for ferrites J. E. Hughes and S. Ratcliffe (*Powder Met.*, 1960, (5), 81–84) A rapid polishing method with diamond on polytetrafluoroethylene laps.

Electrolytic extraction of nonmetallic inclusions from steels by a modified Klinger-Koch method L. Brháček, J. Janáček, and A. Šmrhová (*Hutn. Listy*, 1959, **14**, (k), 54–55) Advantages of the method described compared with the original Klinger–Koch method are that higher current densities can be used, thus shortening the duration of electrolysis; centrally symmetrical dissolution of the sample is obtained; filtration is improved, and manipulation is easier. Operational data are given.

The solubility relations and the precipitation of vanadium nitride in the alpha and gamma field of the system iron–nitrogen–vanadium M. G. Frohberg and H. Graf (*Stahl Eisen*, 1960, **80**, April 14, 539–540) The effect of V on the N in Fe is reviewed in the light of recent publications. The thermodynamic functions of the equilibrium reactions in α- and γ-Fe are derived together with the enthalpies involved It follows that small additions of V can reduce mechanical ageing effects in iron if they are based on the presence of N. Precipitation of VN requires a heat-treatment at high temp. (26 refs).—T.G.

A new nitride precipitate in iron–silicate alloys G. R. Booker and J. Norbury (*Nature*, 1959, **184**, Oct. 24, 1311–1312) A high-purity 3·05%Si alloy was nitrided for 18 h at 640°C, cubic particles appeared which were not identified. After 6 h at 820° the structure became coarser and rod-like particles of α-Si_3N_4 were observed.

Identification of the high-temperature constituent in mild steel surface-hardened by carbo-nitriding M. A. H. Howes, G. T. F. Jay, K. Sachs, and D. Wilkinson (*Nature*, 1959, **184**, Oct. 24, 1312) The structure was close to that of ϵ-Fe_3N but suggested partial replacement by C.

Apparatus for crystallizing alloys at a high cooling rate I. S. Miroshnichenko and I. V. Salli (*Zavods. Lab.*, 1959, (11), 1398–1399) An apparatus is described for preparing thin metal films crystallized with rapid cooling. From experiments carried out a formula has been derived for calculating the cooling rate at which it is possible to achieve a considerable supersaturation of certain solid solutions, for example in Ni–C alloys the α-solid solution of C in Ni was above 1·8% while according to the constitutional diagram maximum solubility at the eutectic temp. is 0·65%C.

Electrical method of measuring recrystallization temperature V. N. Danilov (*Zavods. Lab.*, 1959, (11), 1338–1341) The recrystallization temp. is determined from measurements on deformed metals, of the thermo-emf originating in a closed circuit consisting of a plastically deformed and a non-deformed specimen of the same metal. The recrystallization temp. is read from the maximum of the thermo-emf/temp. curve.

Some features of the recrystallization of transformer steels at high rates of heating. I. Relation between grain size, recrystallization temperature and rate of heating P. D. Izbranov, V. A. Pavlov, and N. M. Rodigin (*Fiz. Met. Metalloved.*, 1959, (6), 915–921) Grain size diminishes with increasing rate of heating but the reduction is small. An increase in the rate of heating by a factor of 10^4 halves grain size. The recrystallization temp. with a high rate of heating, without holding time, is higher than that with a slow rate of heating; but with holding time it is lower.

Cold work and annealing of metals (*Tecn. Indust.*, 1959, **37**, Dec., 1829–1832) The processes occurring in the recrystallization of cold-worked metals are described and a brief discussion of the effects of hot working follows.—P.S.

Mechanism of graphite formation in magnesium iron R. D. Rusev (*Lit. Proizv.*, 1959, (12), 17–19) The two theories current regarding graphite formation: (*a*) direct from the liquid phase, and (*b*) as a result of the decay of the initially formed unstable cementite, are here assessed and an attempt is made at demarcating the influences of each of these processes.

The effect of boron on the recrystallization of silicon iron V. E. Neimark and V. M. Rosenberg (*Fiz. Met. Metalloved.*, 1959, (2), 314–316) The Fe–Si alloy investigated had the composition 3%Si; 0·03%C; 0·2%Mn; 0·01%P. Recrystallization investigation after cold-rolling with 60% reduction was carried out by X-rays. The initial crystallization temp. (for 1 h holding time) changed according to a complex pattern, passing through a maximum with 0·005%B content. Activation energy increased continuously with increasing B content.

The origin of flake graphite formed during high temperature heat treatment of white cast iron W. Oldfield (*BCIRA J. Res. Dev.*, 1959, **7**, Dec., 844–852) An investigation is described to determine whether normal coarse or fine graphite could be obtained by heat treatment of solid white iron. The experiments used an improved system for accurate temp. control of small samples in the region of the freezing range. It is shown that normal flake and undercooled graphite cannot form as the result of decomposition of cementite in the solid state, and that compact or quasi-flake graphite ensues from the decomposition of eutectic carbide in the solid. The results refute the theory that normal flake graphite structures can form in normal cast irons by solidification of white iron which subsequently decomposes in the solid state.—D.L.C.P.

The influence of austenitic grain size on the Ms temperature J. Vero and M. Szikszai (*Acta Techn.*, 1959, **27**, (3–4), 419–427) [In English] The Ms point is usually regarded as dependent upon the composition of the steel but experiments have now been made comparing samples of several steels of different grain sizes. The variation of the temp. at which the transformation begins was examined by dilatometry and by a microscopical method. Doubling the grain size of the austenite raises Ms by 12–13°C.

Room temperature decomposition of austenite in fifty percent nickel–fifty percent iron magnetic alloy tapes N. I. Ananthanarayanan and R. J. Peavler (*J. Appl. Phys.*, 1959, **30**, April, 202s–203s).

Investigating with a magnetometer the austenite transformation under high pressure to martensite A. I. Stregulin and L. A. Mal'nikov (*Zavods. Lab.*, 1959, (11), 1324–1326) The special apparatus consists of a hydraulic press, a magnetometer, and a high pressure chamber. Steel 7N4G4 (0·66%C; 4·2%Ni; 4·74%Mn) was studied and curves were constructed for various temp. and pressures up to 30000 kg/cm². Isothermal transformation can also be studied on this apparatus.

Method of determining residual austenite in surface quenched steel M. M. Klimochkin and S. D. Entin (*Zavods. Lab.*, 1959, (11), 1326–1328) The calculation of residual austenite on the surface quenched layer is based on magnetic measurements and in comparing them, as a reference standard, to a specimen with the least possible amount of residual austenite, formulae for the calculation are deduced.

Heat treating hints: On the formation of martensite E. J. Pavesic (*Met. Treating*, 1959, **10**, Sept.–Oct., 14, 46) An account of the Mf point and its significance.

Effect of neutron irradiation on the martensite transformation in iron–nickel alloys L. F. Porter and G. J. Dienes (*Trans. Met. Soc. AIME*, 1959, **215**, Oct., 854–863) It is shown by electrical resistance and magnetic induction measurements on a 26·4%Ni–Fe and 28·9%Ni-

and 30·4%Ni–Fe alloys that irradiation lowers the Ms temp. and enhances strain recovery on specimens partially transformed to martensite before irradiation. Possible mechanisms are discussed.

Is the martensite transformation isothermal or adiabatic? Z. Nishiyama, A. Tsubaki, H. Suzuki, and Y. Yamada (*Met. Phys.*, 1959, **5,** Nov., 253–254) [In Japanese].

The physical metallurgy of precipitation-hardenable stainless steels D. C. Ludwigson and A. M. Hall (*DMIC Report* 111; *ASTIA AD* 214001; *PB* 151067, 1959, April 20, pp.70) The classification on the basis of austenite stability is noted. The martensitic types transform to martensite on cooling to room temp., which is thought to be highly supersaturated and which precipitate as second phases near 900°F giving additional hardening. The semi-austenitic types remain austenitic on cooling as do the austenitic types. Fabrication is considered and so is susceptibility to intergranular attack by acids.

The effect of inoculation agents on the shrinkage of high alloy chromium and chromium–nickel steels N. S. Kreshchanovskii and M. P. Demin (*Lit. Proizv.*, 1959, (12), 19–21) The linear shrinkage of steel may develop differently and depend on its chemical composition. In the absence of a pearlitic transformation it may be represented by a kinetic curve. In this case shrinkage occurs immediately after the end of pouring and continues to the end of cooling. The shrinkage of steel undergoing a pearlitic transformation consists of a shrinkage period before transformation, a period of expansion (or shrinkage) during transformation and of a period of shrinkage after transformation.

Constitution, metallurgy, and oxidation resistance of iron–chromium–aluminium alloys W. Chubb (*BMI*–1298, 1958, Oct., pp.105; from *US Res. Rep.*, 1959, **31,** March 13, 179, no abstract).

The system zirconium–iron–tin L. E. Tanner (*ARF*–2068–5, 1958, July, pp.61; from *US Techn. Rep.*, 1959, **31,** April 17, 272; no abstract).

Precipitation in a beta-brass–Fe alloy A. E. Berkowitz and P. J. Flanders (*J. Appl. Phys.*, 1959, **30,** April, 111s–112s) A magnetic study of the separation of Fe from β-brass.

CORROSION

The corrosion mechanism of metals. I. The behaviour of the metal/metal oxide electrode in alkali metal halide solutions during potentiometric titration F. Hügli, R. Caperon, P. Bersier, and J. Bersier (*Schweiz. Arch.*, 1960, **26,** Jan., 24–29) The authors explain the method and show the results obtained in the form of graphs.—T.G.

Corrosion mechanisms in the reaction of steel with water and oxygenated solutions at room temperature and 316°C M. B. Strauss and M. C. Bloom (*J. Electrochem. Soc.*, 1960, **107.** Feb., 73–79) Specimens in the form of sealed capsules were used and the pH of the solution was measured. Optical, X-ray, and electron diffraction studies of the corrosion product films were made.

Corrosion kinetics of iron in ammonium nitrate solutions B. Baranowski, T. Ostrowska, and M. Smialowski (*Bull. Acad. Polonaise Sci.*, 1959, **7,** (11), 811–816) [In English] Grain boundary corrosion occurs. The effects of cone of NH_4NO_3 and of NH_4 and NO_3 ions on rate of corrosion are determined. Further work must be carried out to elucidate the mechanism.

Electric fields and corrosion D. Petrocokino (*Corros. Anticorros.*, 1959, **7,** Oct., 335–341) The analogy existing between degrees of potential in the neighbourhood of mine levels and the phenomena accompanying the immersion of steel structures in sea water is discussed.

The corrosion of road vehicle engines by anti-freeze solutions H. H. Collins and R. I. Higgins (*BCIRA J.*, 1959, **7,** June, 667–691) Laboratory and field work carried out by BCIRA are described, and it is shown that corrosion is due to the simultaneous depletion of inhibitors from the solution and oxidation of the glycol to a mixture of organic acids, particularly HCOOH.

Sources of underground corrosion potential differences K. G. Compton (*Corrosion*, 1960, **16,** Feb., 43t–46t) The author discusses the potential of metals in their environments and explains the role of this potential in galvanic corrosion. The sources of potential considered include polarization, stray current, dissimilar metals, differences in electrolyte, and differences in structure surface.—G.F.

Corrosion of metals at Mandapam Camp, India K. S. Rajagopalan, M. Sundaram, and P. L. Annamalai (*Corrosion*, 1959, **15,** Dec., 631t–634t) Monthly rates of corrosion of mild steel, Zn and Cu at two sites at Mandapam Camp on the SE coast of India are reported. Rates of corrosion are discussed in relation to the observed climatic conditions, and differences between open and sheltered tests are compared.—G.F.

Corrosion at the sealing surfaces of flanged joints C. Fischer and H. Zitter (*Werks. Korr.*, 1960, **11,** Jan., 17–22) This type of corrosion is usually caused by penetration of liquid through the seal to the steel, leading to local attack in the absence of inhibitors, and to pitting and stress corrosion as secondary effects, This effect can be largely prevented by the use of non-porous plastic seals. Crevice corrosion, sometimes observed, is usually due to faulty flange construction or sealing.

The corrosion of welded or coated mild-steel and of wrought-iron boiler tubes in highly saline water G. Butler and H. C. K. Ison (*N.E. Coast Inst. Eng. Shipb.*, 1959, Dec. 11, preprint, 95–130) Welded and coated seamless tubes were tested in model boilers. Welded tubes are susceptible to deep pitting and perforation not reduced by increased Cu content. Phosphating and chromating reduce corrosion but not Al treatments. Chromized tubes tended to split on expansion into the tube plates and phosphating appears to have only a temporary effect.

Corrosion in power station feed systems B. Tuck and E. M. Osborn (*Chem. Ind.*, 1960, March 26, 326–331) An account of a trial with *cyclo*hexylamine and morpholine as inhibitors. Both reduced metal pickup of both Fe and Cu, but control of dissolved O_2 is regarded as necessary also.

The residual oil ash corrosion problem C. J. Slunder (*Corrosion*, 1959, **15,** Nov., 601t–606t) The author reviews progress made in combating accelerated high-temp. corrosion caused by combustion products of residual fuel oils. He deals with the mechanism of corrosion, the materials of construction and the use of protective coatings and oil-additives.—G.F.

Corrosion-fatigue in two hot-work die steels D. N. Williams, M. L. Kohn, R. M. Evans, and R. I. Jaffee (*Mod. Cast.*, 1960, **37,** Jan., 19–25) H21 and H23 die steels were teated to determine the relative importance of oxidation and fatigue loading on corrosion-fatigue failures. It was found that sample life in the range 1100–1400°F was relatively independent of oxidation. Variations in heat-treatment, test conditions or other material properties were found to affect sample life in direct proportion to their effect on the plastic strain imposed on the sample during cyclic loading. Sample life could be calculated by use of the equation $N = 1000/S^2$ where N = sample life in cycles. S = % plastic strain imposed on the sample each half-cycle.

The behaviour of stainless steel in the atmosphere A. Noe (*Usine Nouv.*, 1960, **16,** Feb. 25, 29–30) A brief review of reports of tests made in London, and also by the School of Architecture of the University of Princeton, USA, with comparative observations as to the suitability of different classes of stainless steel for use in industrial and country atmospheres.

The most common types of corrosion in stainless steels S. Berg (**Chemische Rundschau*, 1959, **12,** April 1, 185–189) Modes of attack are classified as general corrosion, pitting, intercrystalline corrosion, crevice corrosion, galvanic corrosion, and stress corrosion. These are further considered and examples are given. An all-glass testing apparatus is shown. Much of the data given is derived from the literature.

Can corrosion problems J. J. Daly, jun. (*Corrosion*, 1959, **15,** Nov., 100–106) The author discusses various types of internal and external corrosion encountered in cans. Corrosion control by product modification, container materials selection, packaging methods, storage conditions, protective coatings, and chemical treatment, is considered.—G.F.

Corrosion in a hydrocarbon conversion system F. A. Prange (*Corrosion*, 1959, **15,** Dec., 619t–621t) In a fixed-bed catalytic system for dehydrogenation of butane, metal loss contaminates the catalyst with fine oxide dust which causes undesirable side effects. The way in which the metal loss occurs is discussed, and the effects of alloying agents on both laboratory tests and actual equipment are mentioned.—G.F.

Investigation of cracks in an 18/13/1 stainless-steel autoclave F. Turner and H. K. Richardson (19 *R–TN/C*–1015, 1958, pp.11; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1202) A failure near a weld occurred and a slag deposit was found in the weld root. The slag contained chloride and was connected with the interior by the interface between base plug and wall.

The corrosion resistance of some Mo stainless steels in water hydrolysis of wood and in sulphate cellulose digestion E. Uusitalo and J. Mattila (*Tekn. Kem. Aikak.*, 1959, **16,** (19), 649, 651–652) [In English] Loss of weight of Specimens V 4AE, 4 V4AP, and V 14AS steels were determined in the liquors of which analyses are shown.

Corrosion in the petroleum industry P. Amgwerd (*Rev. Univ. Mines*, 1960, **16,** Feb., 97–101) A brief reference to corrosion problems in the production, stocking, and transport of crude petroleum is followed by a more detailed account of corrosion in refining operations, and of steps taken to minimize its effects.

Attack on metals by synthesis gas from methane-oxygen combustion F. Eberle and R. D. Wylie (*Corrosion*, 1959, **15,** Dec., 622t–626t) Severe corrosion was experienced on 18/8 Nb and 25/20 Cr–Ni steel components of a gas generator used for CH_4–O_2 combustion. The authors describe their metallurgical examination of corroded parts, and discuss the mechanism of corrosion in the different sections of the unit.—G.F.

Accumulated residues attack chemical distilling column (*Corrosion*, 1959, **15,** Nov., 99) Industrial problems caused by residues and deposits are briefly discussed, and data are presented on a cast-iron distilling column which was severely affected by accumulated residue, causing excessive downtime and maintenance.—G.F.

Resistance of materials to corrosion in the heavy chemical industry J. Bigeon (*Rev. Univ. Mines*, 1960, **16,** Feb., 37–54) A review of materials used or considered for equipment used in the chemical industry, in which these materials are classified and the effect of various corrosive agents considered in detail.

Corrosion and decarburization of the ferritic chromium–MOLybdenum steels in sodium-coolant systems W. C. Hayes and O. C. Shepard (*NAA–SR*–2973, 1958, Dec., pp.32; from *US Res. Rep.*, 1959, **31,** April 17, 273; no abstract).

Fabrication of iron–aluminium–chromium alloy fuel plates B. King, J. Neri, and F. G. Tate (*Nuclear Met.*, 1958, **5,** 63–68) Properties of Fe–Al alloys and their corrosion resistance are given as well as studies on fabrication and bonding to itself and to cermets.

Investigation of the iron and steel structures by high-temperature oxidation in air B. I. Kostetskii and O. I. Trotsik (*Fiz. Met. Metalloved.*, 1959, (6), 899–901) This work aimed at comparing the data obtained by previous investigators regarding the microstructure of the oxide phase with the corresponding data regarding the phase observed on the surface of iron due to oxidizing friction. Armco iron and St 45 steel were the materials of the specimens oxidized 2 h at between 650 and 800°. Some of the specimens were cooled in air, others in water. The findings show sharp changes in the structures due to oxygen diffusion. An hypothesis is advanced concerning the oxidation mechanism during friction, of the formation of a solid solution of oxygen in iron under the given cooling conditions.

The effect of heating conditions and alloy composition on the internal oxidation of iron alloys H. Schenck, E. Schmidtmann, and H. Müller (*Arch. Eisenh.*, 1960, **31,** Feb., 121–128) The internal oxidation of Fe–Si, Fe–Al, Fe–Cr, and Fe–Mn alloys was studied in the temp. range 820–1056°C and the effect of heating time, temp., alloy element level, and gas atmosphere composition traced. The results are presented in many graphs and tables. Internal oxidation of all the alloys used increased at a parabolic rate with time and with partial oxygen pressure in the gas atmosphere (43 refs).—T.G.

Effect of silicon on high temperature oxidation of stainless steels J. F. Radavich (*Corrosion*, 1959, **15,** Nov., 613t–617t) In a series of 16%Cr, 10%Ni stainless steels with from 0·17% to 3·55%Si, the author has studied the growth of oxide films at 600°, 800°, and 1000°C using optical, electron microscope, electron diffraction, X-ray diffraction, and X-ray fluorescence techniques. The results show that oxidation resistance increases rapidly as the Si content increases, and the mechanism by which this increased resistance is achieved is discussed.—G.F.

Iron–aluminium alloy systems: II. Comparative study of oxidation resistance in the Fe–Al system R. E. Heimerdinger (*WADC–TR–57–298* (Part II); *AD–207907*; *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1213) Studies of alloys with various contents of Al (0·16)% at up to 2200° and up to 500 h are described. Cr had a beneficial effect only at 7·5%, above and below the effects were deleterious.

Constitution, metallurgy and oxidation resistance of iron–chromium–aluminium alloys W. Chubb, S. Alfant, A. A. Bauer, E. J. Jablonowski, F. R. Shober, and R. F. Dickerson (*BMI–1298*, 1958, Oct. 6, pp.105; from *Nucl. Sci. Abs.*, 1959, **13,** March 15, 389) Ternary sections are shown. The limits of $\alpha+\sigma$ extend from 33–63%Cr at 600°C and decrease at 750°C to 38–57 (at.)%. Limits for σ- and for α-solid solution, the $AlCr_2$ phase etc. are given and the Fe–25–35%Cr–3–8%Al alloys, their properties, fabrication, and casting and the oxidation resistance of Fe–Cr–Al and Ni–Cr–Nb under various conditions are measured.

The adherence of oxide films on metals R. F. Tylecote (*JISI*, 1960, **195,** Aug., 380–385) [This issue].

High-temperature, high-pressure oxidation of alloys caused by carbon dioxide J. C. Bokros and W. P. Wallace (*Corrosion*, 1960, **16,** Feb., 73t–78t) The authors describe the oxidation behaviour of 26 heat-resistant alloys in CO^2 atmospheres in the pressure range 1000–2000 lb/in², and the temp. range 1100–1740°F. The corrosion data are presented, and the effects of time and temp. are discussed.

High temperature metal deterioration in atmospheres containing carbon monoxide and hydrogen W. B. Hoyt and R. H. Caughey (*Corrosion*, 1959, **15,** Dec., 627t–630t) Carbon and stainless steel exposed to gas mixtures rich in H_2 and CO at temperatures of 800–1200°F have been attacked with resulting loss of metal. The authors discuss the conditions under which this attack has been observed in coal-to-oil synthesis plant, and describe the nature of metal deterioration. A possible gas-metal reaction mechanism is postulated.—G.F.

Anodic passivation studies J. D. Sudbury, O. L. Riggs jun., and D. A. Shock (*Corrosion*, 1960, **16,** Feb., 47t–54t) After briefly reviewing the present theories of passivation, the authors describe experimental work which shows that passivity induced by anodic polarization is possible in a wide variety of oxidizing systems. They describe the development of an anodic protection system with high current capacity which makes anodic passivation practical. Laboratory evaluations of any particular system can be extrapolated to field performance with reasonable confidence.—G.F.

Some fundamental features of mechano-chemical attack on metals T. P. Hoar (*AGARD–158*, 1957, Nov., pp.8; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1200) The behaviour of oxide films under chemical attack, fretting, fatigue, and static stress conditions is discussed. Metal yielding under mechanical stress is in a chemically active stage. The mechanism of stress-corrosion cracking is considered.

Equivalent electrical circuit analogy of structure-to-soil potentials R. O. Dean (*Corrosion*, 1959, **15,** Nov., 576t–580t) The author submits concepts and examples of practical applications of the 'equivalent electrical circuit', which is shown to be helpful in understanding and analysing data obtained from structure-to-soil potential measurements.—G.F.

Instruments for measurements in underground corrosion work K. G. Compton (*Corrosion*, 1959, **15,** Nov., 571t–575t) Instruments commonly used for measuring electrical currents and potentials associated with corrosion of underground structures are described and their circuits explained. The measurements considered include voltage, current, resistivity and resistance, and special measurements such as polarization currents.—G.F.

Preliminary report on evaluation of carbon steel H. F. Beeghly (*WAPD–LSR(M)–11*; 1955, May, pp.38; *Nucl. Sci. Abs.*, 1960, **14,** Jan. 31, 237) Loop corrosion tests on A212 steel showed that the rate becomes low after an initial conditioning period. Conditions from pH7 to pH10 with or without H_2 produced very much the same corrosion rate. Effects of Mn and other elements were recorded and welds were also made and tested. Uses of C steels for reactor design are considered.

Some results of corrosion tests with 4% chrome steel E. Herzog (*Corros. Anticorros.*, 1959, **7,** Sept., 312–316) A series of tests with tanks at Kure Beach, N.C., USA, carried out by M. LaQue of International Nickel Co. since 1953 are tabulated and cathodic protection is shown to be generally more effective on moderately alloyed as against unalloyed steels.

Relationship of pH to aerobic corrosion fatigue life of steel F. J. Radd, L. H. Crowder, and L. H. Wolfe (*Nature*, 1959, **284,** Dec. 26, 2008–2009) Studies of AISI 1036 steel are reported in 3%NaCl with NaOH up to saturation. Passivation conditions and the solubility of ferrous hydroxide are discussed.

Factors influencing corrosion G. V. Akimov (*Corrosion*, 1959, **15,** Sept., 455t–468t) A translation of chapter 5 of a book entitled '*Theory of and Research on Metallic Corrosion*'. External and internal factors influencing the rate and distribution of corrosion are discussed. External factors include pH, inhibitors, or stimulators, the concentration of neutral salts, temperature, and rate of solution movement. Corrosion damage due to stray electric currents is fully considered together with methods of mitigation. Internal factors discussed include inherent chemical properties of the elements—the modification of these due to possible protective films is discussed—internal stresses and corrosion fatigue, surface condition, stability of alloys, and the influence of grain size on corrosion. Experimental data are presented on stability limits of solid solutions, while a theoretical discussion on this and related phenomena such as dezincification is also given.

Effects of velocity on corrosion H. R. Copson (*Corrosion*, 1960, **16,** Feb., 86t–92t) The author discusses the effects of the velocity of fluids on the corrosion rate of iron and other metals, and also considers the galvanic effects produced by changes in velocity. Erosion and cavitation effects are explained.—G.F.

Effect of the methods used in the making and working of steel on the pitting corrosion of shipbuilding steel G. Becker and W. Dick (*Stahl Eisen*, 1960, **80,** April 14, 491–497) Pitting of shipbuilding steel is affected to a much greater extent by the surface quality of the steel than by its composition. From the tests carried out it is evident that the plates for shipbuilding should be thoroughly descaled and a protective coating applied. For economic reasons such an operation should be carried out in the rolling mill. Means of avoiding pitting in tankers are briefly mentioned, including cathodic protection, limitation of the temp. of the cleaning water, etc. (21 refs).—T.G.

Comparative corrosion tests on chromium-plated articles W. Nohse (*Draht*, 1960, **11,** March, 112–113) A résumé of a lecture delivered at a symposium on electroplating held at Schwäbisch-Gmünd in Sept.–Oct. 1959. Corrodcote, artificial industrial atmospheres containing CO_2 and SO_2, and salt-spray tests were used for an evaluation of the corrosion-resistance of Ni–Cr plated articles. It was found that protection is provided only if both the Ni and the Cr coatings are pore-free.—T.G.

Corrosion of nickel-coated mild steel B. C. Banerjee (*J. Electrochem. Soc.*, 1960, **107,** Feb., 80–82) Mild steel discs coated with Ni with (100), $(10\bar{1}0)+(211)$, (110) and (210) orientations in HCl showed the last-named to be most resistant. In NaCl no appreciable differences were found. Mechanisms are discussed.

Corrosion data for chemical nickel plating J. R. Spraul (*Plating*, 1959, **46,** Dec., 1364–1369) An account of the process and an outside exposure test programme is given with an extensive table of resistance of plated panels to a variety of chemicals and solutions.

Theoretical and practical problems concerning the use of cans made from tinplate P. Nehring (*Mitt. Forsch. Blechverarb.*, 1959, July 15, 197–204) Underlying principles governing tinplate corrosion are explained and their practical implications discussed.

The corrosion resistance of, and an evaluating method for, various electrodeposits on steel H. W. Dettner (*Plating*, 1959, **46,** May, 469–480) Tests are reported which compared the corrosion resistance of electro-plated Zn, Cd, Sn, and Sn–Zn alloy deposits in thicknesses of 2, 5, and 10 μ by means of salt spray, humidity, SO_2–CO_2, and outdoor exposure tests. Zinc deposits, especially when chromated, showed in general, a better behaviour than tin, except acid tin in salt spray and industrial air tests. The behaviour of Sn–Zn deposits was generally between that of Sn and Zn except in the salt spray test where the alloy-coated steel rusted at a very slow rate.—D.L.C.P.

Additional studies with an all-glass multiple-test apparatus for nitric acid testing of stainless steels G. F. Tisinai and C. H. Samans (*ASTM Bull.*, 1959, (238), 64) Notes on work with this equipment on the corrosion testing of stainless steels (*ibid*,. 1956, (214), May, 57, for a description of the apparatus).

Internal boiler water treatment C. Hinchcliffe (*Australasian Eng.*, 1959, **51,** April 7, 57–60) The development and application of an internal boiler water treatment involving control of Mg content is described. The investigation work showed that a minimum Mg content in the boiler water was necessary to ensure a 'mobile' boiler sludge. In practice, scale has been prevented by increasing the Mg content of the boiler water as necessary.

Dissolution of zirconium and stainless steel corrosion product oxides J. W. Snider and P. A. Haas (*CF–58–8–28*, 1958, Aug., pp.16; from *US Res. Rep.*, 1959, **31,** April 17, 273; no abstract).

Cavitation erosion: Comparative tests R. Mathieson and J. M. Hobbs (*Engineering*, 1960, **189,** Jan. 22, 136–137) A review and comparison of drop-impact and flow tests.

The corrosion of 18/13/1 steel under conditions simulating those in the highly active liquor evaporators at Windscale H. N. Jones (*RDB(C)/TN–27*, 1953, April, pp.10; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1206) Tests at 70, 80, and 100°C were carried out; the corrosion was much less at lower temp.

Fundamental study of the corrosion of 18/13/1 chromium nickel steel in nitric acid H. N. Jones (*RDB(C)/TN–44*, 1953, June, pp.17; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1206) Attack takes place at grain boundaries where possibly a phase exists which encourages attack. This could not be identified. Alloys containing Nb should be examined for boundary attack.

The effects of various contaminants on the corrosion of 18/13/1 chromium–nickel steel in boiling 70% nitric acid H. N. Jones (*RDB(C)/TN–25*, 1953, April, pp.9; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1206) Additions of CrO_4-ion increase the rate. V accelerates corrosion and Ce and Mo to a lesser extent, W increases the rate only slightly and Co seems to have no effect.

The influence of various ions on the rate of corrosion of 18%Cr–13%Ni, Nb-stabilised austenitic steel in 70% w/w nitric acid T. E. Evans and H. N. Jones (*RDB(C)/TN–75*,

1954, June, pp.19; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1207) Ce, V, Mo, and Th were tested. The effects are discussed.

Peening cuts stress corrosion H. Suss (*Steel*, 1959, **145,** Dec. 21, 88) Results of tests are reported, which suggest that peening of stainless steel may retard or eliminate cracking effects of high temp. water.—D.L.C.P.

The influence of arsenic on the hydrogen penetration into iron cathodes polarized in sulphuric acid solutions H. Angerstein-Kozlowska (*Bull. Acad. Polonaise Sci.*, 1959, **7,** (12), 881–886).

Sodium silicate as a stress corrosion inhibitor M. J. McGoff (*MSAR*–59–113; 1959, Oct., pp.18; from *Nucl. Sci. Abs.*, 1960, **14,** Jan. 31, 240) Sodium silicate was used with insulating materials in tests against chloride stress corrosion of type 304 pipes.

Reduction of oil-ash corrosion by use of additives in residual fuels H. W. Schab and F. R. Gessner, jun. (*EES rept.* 070034C; *PB* 137101, 1957, April, pp.55; from *US Res. Rep.*, 1959, **31,** March 13, 167) Tests on AISI 310 with high-V fuel at 1500–1700°F in small combustion rigs were carried out with 34 additives, La, Sb, and Fe were most promising and Ca, Na, Mn, rare earths, Ce, and Nd also showed good qualities.

Action of the XO_4^{n-} inhibitors G. H. Cartledge (*Corrosion*, 1959, **15,** Sept., 469t–476t) Experimental studies of the effect of ions of the type XI_4^{n-} on the corrosion of iron and steel are summarized. The local corrosion current density is expressed in terms of the activation energies of the postulated activated complexes that act as intermediaries and of their distribution over the entire surface. Studies have shown that the inhibited iron surface is in an effectively static condition but that a dynamic, reversible state exists between the surface film and the solution. Pertechnetate ion has completely prevented corrosion of iron in an aerated solution for five years. Passivation by OsO_4 is so complete that the iron electrode has the potential of the Os^{IV}–Os^{VIII} couple. The Flade potential can be observed on aerated inhibited iron electrodes indicating electrochemical identity of the films formed with those produced by more vigorous passivation. It is believed that the films are mainly due to oxygen. The role of the adsorbed inhibitor, in so modifying the kinetics of the electrode processes at the metal/solution interface as to produce passivation is discussed. Differences in the ability of various XI_4^{n-} ions to produce inhibition may be due to the nature of the electrostatic field exerted by them over short distances.—J.F.S.

Control tests for filming amines K. B. Coates (*Corros. Techn.*, 1960, **7,** Feb., 46) The filming amine, such as Afloc Dispersion No.190, is detected by the pink coloration formed with cosine, and can be titrated with an amionic surface-active agent, sodium dioctyl sulphosuccinate solution being used in the presence of a buffer.

A proposed standardized static laboratory screening test for materials to be used as inhibitors in sour oil and gas wells NACE Technical Unit Committee T–1K (*Corrosion*, 1960, **16,** Feb., 63t–64t) A description is given of a static laboratory test designed to provide a simple, easily reproducible method for screening materials to be evaluated subsequently as possible inhibitors in sour oil and gas wells. A standard coupon is immersed in a standard corrodent as a control, and in an identical system a definite concentration of the test material is added; the criterion of success is the amount of weight loss suppression afforded by the material.—G.F.

Electrical resistance method for studying corrosion inhibitors in automotive anti-freeze J. C. Cessna (*Corrosion*, 1959, **15,** Nov , 607t–612t) The author discusses the application of the resistance method to the determination of relative corrosion inhibitor effectiveness in the aqueous ethylene glycol system The corrosion of Fe, Al, and Cu alloys is reported for various test conditions, and the effect of certain corrosion inhibitors is illustrated.—G.F.

Corrosion inhibitors for carbon tetrachloride-water vapour atmospheres K. W. Calkins and R. W. Hawley (*Corrosion*, 1959, **15,** Sept., 447t–449t) Of 48 inhibitors tested, ethanolamine was the best for preventing the corrosion of steel in a CCl_4–H_2O vapour system.

Screening tests of inhibitors to prevent chloride stress corrosion J. H. Phillips and W. J. Singley (*Corrosion*, 1959, **15,** Sept., 450t–454t) Screening tests have been conducted in a tilting autoclave of a wide range of possible inhibitors for austenitic stainless steel for use in alkaline-phosphate boiler water containing 50– to 500–ppm chloride. Nitrate appeared to be a suitable inhibitor. Na_2SO_3, by scavenging O_2, also prevented stress corrosion cracking in the autoclave tests.—J.F.S.

Corrosion testing of aerosol products M. J. Root and M. J. Maury (*Corros. Techn.*, 1959, **7,** March, 85–87) Accelerated electrolytic tests for testing the corrosion resistance of cosmetics and other aerosols to be stored in tinplate containers are described.

Interactions between four volatile corrosion inhibitors R. L. Le Mar (*PB* 151010, 1957, Dec., pp.43; from *US Res. Rep.*, 1959, **31,** March 13, 167) VCI on kraft papers were tested on steel panels.

The Flade potential of iron passivated by various inorganic corrosion inhibitors H. H. Uhlig and P. F. King (*PB* 140627, 1956, pp.26; from *US Res. Rep.*, 1959, **32,** Aug. 14, 238) Passivation by ferrites, nitrites, chromates, molybdates, tungstates, HNO_3, and anodic polarization were studied by measuring the potential just before decay of passivity.

Methods of preventing the corrosion of iron and steel D. Iturrioz (*Met. Elect.*, 1958, **22,** Aug., 65–70; Sept., 99–105) [In Spanish] The mechanism of corrosion is examined with reference to methods of protection which are described in detail.—P.S.

Possible uses of zinc in the struggle against marine corrosion R. Souské (*Corros. Anticorros.*, 1959, **7,** Oct., 356–359) Different methods of application of a coating of Zn to the surface of steel as a protection against marine corrosion are examined; electrolysis, metallization, sherardizing, Zn paint, and cathodic protection are described and discussed.

Corrosion of steel: Part III. Protective behaviour of some paints A. K. Choudhury and S. C. Shome (*J. Sci. Ind. Res.*, 1959, **18A,** Dec., 568–570) Immersion tests in salt solution with single pigments in raw linseed oil are reported. Zn chromate and molybdate give good protection. Zn tungstate is slightly inferior. Pb paints gave good protection at first but gradually broke down on continued immersion.

The use of products of coal distillation for protection against corrosion C. Hanfland (*Bol. Inf. Inst. Nac. Carbon*, 1959, **8,** May–Aug., 93–110) A general account of corrosion problems, especially the effects of lamination and rolling scale on painted iron and steel. The use of hot tar or cold tar compositions is described and the more recent synoplasts and their value even in the worst conditions. The materials are adequate for mining conditions.

Some methods of controlling hydrolytic deposits P. Godeaux (*Usine Nour.*, 1959, **16,** Nov. 12, 97–99; *Métaux*, 1958, Dec., 519) Methods of control are described under the following headings: examination of appearance of deposit (1 method); control of adherence (4 methods); determination of thickness of deposit (2 methods); control of porosity, and resistance to corrosion (5 methods).—S.H.–S.

Combating corrosion in molten extraction processes H. Susskind, F. B. Hill, L. Green, S. Kallsh, L. Kukacka, W. E. McNulty, and E. Wirsing (*Chem. Eng. Prog.*, 1960, **56,** March, 57–63) Tests on dilute solutions of metals in Bi are reported. Steels and non-ferrous metals were used.

Conditioning treatment for acid cleaned cooling system W. Rosenfeld (*PB* 151527, 1958, Dec., pp.10) Protection against corrosion by antifreeze agents by cleaning at proper intervals is described.

Application of anodic protection in the chemical industry D. A. Shock, O. L. Riggs, jun. and J. D. Sudbury (*Corrosion*, 1960, **16,** Feb., 55t–58t) The authors discuss the important factors in the evaluation of a given metal–electrolyte system for the application of anodic protection, quoting a number of examples. The applicability and economic aspects of the anodic protection process are considered, and the effects of temperature and type of alloy are discussed.—G.F.

Anodic control of corrosion in a sulfonation plant O. L. Riggs, jun., M. Hutchison, N. L. Conger (*Corrosion*, 1960, **16,** Feb., 58t–62t) A newly developed automatic potential controller has been in continuous use for a year in controlling corrosion in a mild steel oleum storage tank and in a stainless steel sulphonic acid neutralization tank. The authors describe the preliminary laboratory tests and the application of the control system, and quote results illustrating its effectiveness.—G.F.

Corrosion prevention in automobile bodies T. P. Hoar (*Inst. Mech. Eng. Proc. Auto. Div.*, 1957–58, (2), 53–73) Preparation, phosphating, and finishing are reviewed and design and testing are discussed. Plating, Al coating, and the use of stainless steel are also considered and the effects of storage, cleaning, and polishing are added.

Studies on the susceptibility of cathodically protected steel to hydrogen embrittlement W. H. Bruckner and K. M. Myles (*Corrosion*, 1959, **15,** Nov., 591t–595t) The authors discuss the results of tests initiated to determine whether steel structures buried in soil and placed under cathodic protection could be embrittled by H_2 as a consequence of the cathodic protection. The conditions under which embrittlement can occur are outlined.—G.F.

Cathodic protection of Southern Pacific high-pressure products pipe line D. N. Miller (*Corrosion*, 1959, **15,** Dec., 104–106) The methods used to achieve cathodic protection of a pipe line across central California and Nevada are discussed. Special problems encountered are outlined.—G.F.

The use of Ti and Ti alloy anodes for cathodic protection with independent current supply H. Richaud (*Corros. Anticorros.*, 1959, **7,** Sept., 322–324) The uses of Ti and Ti alloy anodes and their advantages compared with anodes of steel, graphite, platinum, etc., are described.

The problem of cathodic protection of steel immersed in sea water A. Hache (*Corros. Anticorros.*, 1959, **7,** Sept., 307–312) Reasons for locating the Sea Water Corrosion Testing Station at Biarritz are explained. A comparison is made between test pieces placed in a tank of circulating sea water with cathodic protection and others placed in open sea water, and the results are described.—S.H.–S.

Cathodic protection its application in the marine field B. Ravelot (*Corros. Anticorros.*, 1959, **7,** Oct., 342–355) After examining the theories of states of immunity, passivity, and corrosion, the author discusses the possibilities of cathodic protection and the uses of two processes are described.—S.H.–S.

Protecting the 'Landaura' (*Corr. Prev.*, 1959, **6,** Sept., 41) Treatment of the bottom plates of a 12-year-old vessel is briefly described.

Some platinum anode designs for cathodic protection of active ships H. S. Preiser and B. H. Tytell (*Corrosion*, 1959, **15,** Nov., 596t–600t) The authors present voltage–current relationships which have been determined for Pt wires, gauzes, and discs, and a Pt-clad–Ta disc, in laboratory and harbour sea-water tests. The Pt-clad Ta disc produced the best current–voltage ratio.—G.F.

Lead–Silver alloy anodes in cathodic protection of ships K. N. Barnard, G. L. Christie, and D. G. Gage (*Corrosion*, 1959, **15,** Nov., 581t–586t) A Pb-2%Ag alloy has been used experimentally as an impressed current system for cathodic protection of an active destroyer (H.M.C.S. *Crusader*). The authors describe the anodes and their associated equipment and method of fitting. Data from one year's service are given, and a comparison of performance and costs is made with existing impressed current and galvanic systems.—G.F.

Electrical bonding of cathodically protected to unprotected ships G. A. Bennett, N. S. Dempster, and A. J. Wallace (*Corrosion*, 1959, **15,** Nov., 587t–590t) The authors show that serious corrosion of the hull of an unprotected ship berthed alongside a cathodically protected vessel can be avoided by low-resistance (<0·0005 ohm) electrical bonding of the hulls.—G.F.

Ultrasonic measurement of tank ship corrosion losses D. J. Evans (*Corrosion*, 1959, **15**, Nov., 561t–565t) The author describes an ultrasonic method of thickness measurement from one side of a steel plate, and gives details of the 'slide rule' method of calibration. Sources of error are discussed, and means of overcoming them are considered.—G.F.

Controlling internal corrosion of tank ships J. F. Koehler (*Corrosion*, 1959, **15**, Nov., 557t–560t) The author describes the measures taken by the Esso Standard Oil Co. to control internal corrosion in tankers. They include partial and complete cathodic protection, inhibition, coatings, and combinations of these methods. The relative advantages and disadvantages of the different methods, and their economic aspects are discussed.—G.F.

Anti-corrosion treatment for Thames barge (*Corros. Prev.*, 1959, **6**, June, 56) The protective coatings applied to a recently commissioned tug are enumerated in detail, together with a special engine cooling water system designed to prevent corrosive action by the river water.

'Copon' coatings for marine corrosion control K. H. Reynolds (*Corros. Prev.*, 1956, **6**, June, 57–59) The results of several years' successful trials in the protection of offshore drilling rigs and equipment, and ship's hulls by an activated *Copon* coating are given. The composition is not disclosed.

Use of high silicon cast iron for anodes Technical Unit Committee T–2B of NACE (*Corrosion*, 1960, **16**, Feb., 65t–69t) Data are tabulated from questionnaire replies received from more than 100 companies pertaining to the use of high silicon cast iron for anodes. In ground bed applications, the test anodes are at least equal to graphite anodes; in fresh water applications, they perform very well if the temp. does not exceed 125°F; but at high temp. or in salt water applications where the temp. may exceed 100°F, a Mo-containing high-Si iron is to be preferred.—G.F.

ANALYSIS

Determination of gases in metals and alloys by vacuum techniques N. G. Banerjee (*NML Tech. J.*, 1959, **1**, Nov., 28–33) The author reports his impressions in this field obtained during a tour in the United Kingdom.

Vacuum fusion provides quick and accurate gas analyses of metals V. S. McMullen (*Blast Furn. Steel Plant*, 1960, **48**, Feb., 188) A description of the operation of the Serfass–Fischer equipment at the works of Vanadium Corporation of America.

The rate-determining step in desorption of gases from solid specimens B. Baranowski (*Bull. Acad. Polonaise Sci.*, 1959, **7**, (12), 887–890) Later papers study desorption from metal films.

Application of the Dumas method to the determination of traces of nitrogen in iron and steel J. Coulombeau and E. Jaudon (*Chim. Anal.*, 1960, **42**, Feb., 61–68) The sample is heated with lead borate at 1050°C in a stream of CO_2, the evolved gases are purified over copper oxide, hopcalite and I_2O_5 and N_2 is measured over 50%KOH.

Development and application of an electrochemical method of measuring oxygen K. Damaschke (*Materialprüfung*, 1959, **1**, June, 209–212) The method described can be used for measuring changes in concentration of O_2, and various applications are described. The method can be used both for oxidizing and for reducing substances.

Determination of oxygen in iron and steel M. L. Pearce and C. R. Masson (*Nature*, 1960, **185**, Feb. 6, 373–374) Studies by the method of isotope dilution show values consistently higher than vacuum-fusion data.

The determination of sulphur in cast iron H. Green (*Metallurgia*, 1959, **60**, Nov., 229–232) Possible sources of loss in the determination of S by the combustion method are examined. For even approximately quantitative recovery SO_3 must be retained and the determination must be completed by an alkali titration; V_2O_5 would be a useful flux under these conditions.

A radiochemical investigation of the combustion method for the determination of sulphur in cast iron R. C. Rooney and F. Scott (*JISI*, 1960, **195**, Aug., 417–421) [This issue].

Separation of some heavy metals by solvent extraction and by paper chromatography G. Weidmann (*Canad. J. Chem.*, 1960, **38**, March, 459–464) Work on Fe, Co, and Cu shows that they can be extracted from Ni and further separation can be effected by change of acid concentration using tributyl phosphate.

On possible analytical applications of some aromatic oximes. I. 2-carboxy-*iso*nitrosoacetanilide and its reactions with metallic ions F. Buscaróns and L. M. Munné (*Anal. Real. Soc. Esp. Fis. Quim.*, 1959, **55**, (B), Dec., 789–794) The preparation of this compound and its methyl ester is first described and a study of the qualitative reactions of 45 different metallic ions in acid, neutral or alkaline solutions with these compounds is then reported. The selectivity of the $-CO-\overset{1}{C}=NOH$ group is shown by the fact that only Pd reacts in an acid solution (sensitivity 1:200000), and Cu, Fe, Co, and Ni in alkaline solutions, with both reagents. The latter result offers a method for the colorimetric determination of Co since the other elements either precipitate out or can be complexed with cyanides to colourless forms. The sensitivity is 1:1000000 and Beer's Law is obeyed even at concentrations of 1–2 γ/ml.

The extraction of ferric chloride with methyl *iso*butyl ketone and amyl acetate A. Claassen and L. Bastings (*Z. Anal. Chem.*, 1958, (6), 403–409).

The absorptiometric determination of iron in uranium plant solutions, and plant intermediates (2:2′-dipyridyl method) (*IGO–AM/W–184*, 1958, pp.6; from *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1164).

Analytical applications of the chelones. XXII. Direct volumetric macro- and micro-determination of nickel with EGTA, using Murexide as end-point indicator F. Bermejo Martinez and M. Paz Castro (*Anal. Real. Soc. Espan. Fis. Quim.*, 1959, **55(B)**, Dec., 803–806) A method using ethylene glycol bis-(β-aminoethyl ether)-*N*:*N*′-tetra-acetic acid, has been found to give very accurate results in determining the nickel in prepared solutions (of 100 ml) ranging from 0·328 to 131·2 mg. **XXIII. Direct volumetric micro-determination of cobalt with EGTA, using Murexide as end-point indicator** (807–808) The method proposed, using the same reagent has been found to give good results for Co levels ranging from 0·304 to 7 mgs but with greater concentrations the errors become too large.—F.S.

Nitroso-R-Salt: reagent for cobalt and iron R. J. Reynolds and P. Casapieri (*Org. Reagents for Met. Monograph No.*33, 1957, pp.7) (57 refs).

The use of the polarograph in ferrous metallurgical laboratories. I. The determination of copper, nickel and cobalt in presence of one another in steels G. Thanheiser and G. Massen (*SLA–59–20473*, trans. from *Trans. Kaiser–Wilhelm–Institut Eisenforsch. Mitt.*, 1937, **19**, (2), 27–46; from *US Techn. Trans.*, 1960, **3**, Jan. 8; no abstract).

The potentiometric reduction of solutions containing molybdenum and tungsten ions H. K. El-Shamy and M. F. Barakat (*Egyptian J. Chem.*, 1959, **2**, 191–205).

Separation and gravimetric determination of niobium, tantalum and titanium by precipitation with *N*-benzoyl-*N*-phenylhydroxylamine F. J. Langmyhr and T. Hongslo (*Anal. Chim. Acta.*, 1960, **22**, April, 301–305) The method was applied to analysis of minerals.

Formal oxidation potentials of iron, plutonium and neptunium determined with a controlled potential coulometer R. W. Stromatt, R. M. Peekema, and F. A. Scott (*HW–58212*, 1958, Nov. 12, pp.23; from *Nucl. Sci. Abs.*, 1959, **13**, May 31, 1164). The Fe (II/III) potential was determined as −0·767 volts.

Development of apparatus and methods for chemical and electrochemical spot testing titanium in stainless steels (*PB* 139162, 1957, Oct., pp.3; from *US Res. Rep.*, 1959, **32**, Dec. 11, 682) Chromotropic acid is used.

The analytical chemistry of zirconium G. W. C. Milner and J. W. Edwards (*Analyst*, 1960, **85**, Feb., 86–97) A review (139 refs).

Improvement of accuracy in the spectrochemical analysis of high alloy steels Res. Inst. of Iron and Steel, Peking (*Acta. Phys. Sin.*, 1959, **15**, June, 285–295) [In Chinese] It is shown that, with refinement of all stages of analysis, the min. standard deviation for the photographic method of spectrochemical analysis of high-alloy steels is about 1%.

Internal standard method of general spectrographic analysis F. M. Smith (*HW–59864*; 1959, April, pp.15; from *Nucl. Sci. Abs.*, 1959, **13**, Sept. 15, 2020) A moderately precise method for 20 elements with an average deviation of ±15% is described.

Spectrochemical analysis of carbon in steels and cemented layers of cementation steels Chen Bin, Din Li-Tszyun, and Val Lin (*Acta Phys. Sin.*, 1959, (10), 513–520) [In Chinese, with summary in Russian] A method is described of quantitative analysis of carbon in these steels. A quartz spectrograph is used with two sources, a spark generator and a high-frequency generator. The mean-square error is 3·65% for the spark generator and 5·77% for the high frequency sparking.

Rapid analysis of steels on a model of a photoelectric instrument V. N. Vovk and I. G. Savich (*Zavods. Lab.*, 1959, (8), 969–970) A photoelectric apparatus is described for emission spectrographic analysis constructed in 1956 by the State Optical Institute (*Izvest. AN SSR seria Fiz.*, 1955, **19**, 1) which since May 1957 has worked round the clock in a steel plant at a distance of 40–50 m from the furnaces. A constant temp. of 22–24° is maintained in the space surrounding the apparatus. Some results are given.

Infrared absorption spectroscopy, a modern aid to chemical analysis and material testing E. G. Hoffmann (*Elektrowärme*, 1959, **17**, Aug., 310–313) The use of automatic scanning infrared spectrophotometers for scanning infrared absorption spectra is discussed.

Methods of spectrographic analysis of high purity metals L. N. Filimonov (*Zavods. Lab.*, 1959, (8), 936–946) A review of the work of many investigators both in Russia and abroad (67 refs).

Spectrochemical determination of tungsten in the presence of molybdenum N. I. Tarasevich, A. D. Khlystova, and E. A. Pak (*Zavods. Lab.*, 1959, (8), 955–956) A method of analysis is described suitable for small amount of W (of the order of 10^{-3}%) in Mo while by direct spectral analysis it may be determined only down to 0·05–0·02%. A method of chemical enrichment is proposed using inorganic substances such as silicic acid, metastannic acid, zirconium phosphate, and ammonium phospho-molybdate which makes spectrographic analysis possible.

Standards for spectrographic analysis L. N. Filimonov and V. U. Polyakova (*Zavods. Lab.*, 1959, (8), 972–983) Some methods of preparing and testing the standards are given and the determination of the concentration of the elements in the standards is explained.

Spectrochemical analysis of steels by the method of transfer to counter-electrodes A. Garrido de Bartolomé (*Rev. Ciencia Apl.*, 1959, **13**, July–Aug., 322–325) The method is applied to high-alloy steels, using a Hilger high-dispersion spectrograph. A probable error of ±8% and ±10% are obtained for Cr and Ni respectively.

Spectrographic analysis of solutions with spark excitation A. Larsson (*Jernkont. Ann.*, 1959, **143**, (8), 508–522) The method described involves partial evaporation of a drop of solution on the Cu or Ag electrode tip. Details are given of procedures for analysis of steel, Cu alloys, and slags.

Zirconium analysis by production control quantometer C. L. Easterday (*Anal. Chem.*, 1959, **31**, Nov., 1867–1868).

Layer-by-layer analysis of aluminised steel by spectral line-width methods Heat Treat. Res. Inst., Hankow (*Acta. Phys. Sin.*, 1959, **15**, June, 305–310) [In Chinese] The relationship between Al concn. and line width can be used for quant. analysis of the aluminized layers.

The removing of disturbances in the working of spectrophotometers and other measuring instruments caused by spectrographic sources of excitation J. Czakow (*Chemia Anal.*, 1959, **4**, (4), 753) It was found in the department of chemical analysis that spark and arc generat-

ors used in emission spectrographic analysis affected the readings of other measuring instruments. This fault was rectified by introducing additional earthing.

Resorcinol as a reagent for the spectrophotometric determination of ferric iron G. S. Rao (*Anal. Chim. Acta*, 1959, **21**, Dec., 564–568).

Spectrophotometric determination of vanadium in iron and steel by phosphotungstic acid method H. Kitagawa and N. Shibata (*Nippon Kinzoku*, 1958, **22**, Nov., 580–583) An optimum procedure is described, based on masking Fe, Mo, Ti, and other elements by addition of HF.—K.E.J.

Spectrophotometric determination of cobalt after extraction of the thiocyanate complex with acetylacetone W. B. Brown and J. F. Steinbach (*Anal. Chem.*, 1959, **31**, Nov., 1805–1806) Fe and Cu are not extracted under the conditions given.

Spectrophotometric determination of small amounts of lead in cast iron K. Ota and S. Mori (*Nippon Kinzoku*, 1959, **23**, Jan., 42–44) The existing method using dithizone extraction involves warming the soln. with NH_2OH,HCl; Rochelle salt and KCN, for masking Fe^{III}, but gives variable results. This is obviated by reducing Fe^{III} with NH_2OH,HCl at pH 2·5–3·0, and then adding Rochelle salt and KCN. With benzene as solvent for dithizone, one extraction suffices. The method is suitable fo cast iron and carbon steel containing 0·003–0·04%Pb.—K.E.J.

Rapid spectrophotometric determination of vanadium in all materials P. Blanquet (*Chim. Anal.*, 1959, **41**, Sept., 359–367) V up to 1% is determined with an absolute accuracy of ±0·01% in the presence of all proportions of Fe, Co, Ni, Mo, Cr, W, and other materials, by using the blue coloration obtained with pyrocatechol in an ammoniacal medium. Optical density is measured at 800 mμ. A modification is described for V contents > 1%.

Direct-reading spectrochemical analysis using X-ray and optical emission techniques J. H. Kelly (*Can. Min. Met. Bull.*, 1959, **52**, Nov., 687–693) Steel analysis is outlined.

X-Ray fluorescence analysis of stainless steel in aqueous solutions R. W. Jones and R. W. Ashley (*Anal. Chem.*, 1959, **31**, Oct., 1629–1631) Nb, Mo, Cr, and Ni are determined, the three latter after separation of the Nb.

Quantitative analysis of Fischer–Tropsch catalysts by X-ray diffraction. Determination of alpha-iron, magnetite, and iron carbides F. H. Herbstein, J. Smuts, and J. N. van Niekerk (*Anal. Chem.*, 1960, **32**, Jan., 20–24).

Studies on analytical methods for trace elements in metals by using a radioactive isotope. I. Determination of zirconium in steel. II. Determination of niobium and tantalum in steel H. Amano (*Sci. Rep. Res. Inst. Tohoku Univ.*, 1959, **11**, (4), 296–303; 304–311) [In English] I. Using ^{95}Zr as tracer, the accuracies of the phosphate, arsenate, basic selenite, and mandelate methods for determining Zr, and mercury cathode electrolysis and ion-exchange resin methods for separating Zr from iron were examined. Zr should be pptd. as phosphate after separation by mercury cathode electrolysis (14 refs). II. Methods for determination and separation of Nb and Ta were studied, using ^{95}Nb and ^{182}Ta as tracer and radioactive indicator. The best separation method used perchloric acid hydrolysis; after the single separation from Fe, the sum of Nb and Ta was titrated with Cupferron using ^{182}Ta as radioactive indicator (11 refs).—K.E.J.

Studies on analytical methods for trace elements in metals using radioactive isotopes. II. Radiometric titration of niobium and tantalum in steel H. Amano (*Nippon Kinzoku*, 1958, **22**, Dec., 625–628) A successful method is based on the use of ^{95}Nb and ^{182}Ta as tracers. After a single separation from iron using the perchloric acid-hydrolysis method, titration is carried out on a cupferron soln. with ^{182}Ta as radioactive indicator (11 refs).—K.E.J.

INDUSTRIAL USES AND APPLICATIONS

Modern hot-working steels for tools for the hot working of non-ferrous metals H. M. Hiller (*Metall.*, 1960, **14**, April, 310–317) The advantages offered by modern Mo-bearing steels for hot-working tools for non-ferrous metals as compared with the hitherto used W steels are pointed out. In order to help the toolmaker and consumer of tools in the selection of steels a number of examples are quoted in which the two steel types are compared.—T.G.

How to select stainless steel for use in the mining industry. Part II. Experience with stainless in the field R. E. Paret (*Eng. Min. J.*, 1960, **161**, Jan., 80–84) Use of stainless steel for mining application in corrosive and abrasive conditions, in sizing and classification, smelting, and refining is illustrated by examples.

Stainless steel as automotive trim C. O. Durbin (*AIMME, Elect. Furn. Steel Conf. Proc.*, 1958, **16**, 103–105) The author briefly outlines the extent to which stainless steel is used for exterior trimming on automobiles. The influence of colour on the choice of stainless steel type is discussed.—G.F.

Steels in the manufacture of automobile parts W. de Lima e Silva (*Eng. Min. e Met.*, 1959, **30**, Nov., 257–264) [In Portuguese] A range of steels, proposed as standards, is discussed and, for each steel a table of properties is given. These tables give general details, hot working and heat-treatment data, and mechanical property charts. Substitute steels are listed and commented upon. A final table lists the range of parts, their working conditions, the properties required, and the recommended SAE numbers.—P.S.

Materials for exterior bright trim on passenger cars D. N. Frey (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 105–109) The factors governing the choice of material for exterior trimming on automobiles is discussed, reference being made to exterior trim on a typical 1959 car.—G.F.

Survey of exhaust valve materials C. W. Vigor (*AIMME, Elec. Furn. Steel Conf. Proc.*, 1958, **16**, 110–115) The author reviews recent trends in the use of exhaust valve materials and discusses the reasons why better materials are required. The three basic types of stainless steel (ferritic, sigma-hardening, and austenitic) currently available for valve service are considered.—G.F.

Seven uses of precipitation hardenable stainless steels (*Mat. Design Eng.*, 1960, **51**, Jan., 132) The advantages offered by four grades of alloys, 17-4 PH (reactors), 17-7 PH (microswitch interior springs), AM-350 (Wing-web sections, and fuel tank and camera pod of X-7 missile), and AM-355 (fuel container for solid fuel rockets), are stated and their uses illustrated.—S.H.-S.

High strength alloy pipe answer to thermal stress Freeport Sulphur Co. (*Steel*, 1960, **146**, Jan. 18, 97) The arrangement of three concentric Mo–Mn steel pipes forming a 7 mile pipeline for molten sulphur is described. There are no expansion joints or loops. The outer protective pipe is prestressed and fastened to the inner pipes while they are at an intermediate temperature; the stress at all temperatures is thus minimized.—D.L.C.P.

Economic comparison of zircaloy and stainless steel in nuclear power reactors M. Benedict (*NP*–7260, 1958, Feb., pp.35; also in *Energia Nucleare*, 1958, **5**, 493–508; from *Nucl. Sci. Abs.*, 1959, **13**, March 31, 1205) Zircaloy is less expensive than stainless steel and can be used extensively. The balance point in the prices for various reactors was calculated.

Titanium v.s. steel—a symposium. Introduction W. J. Harris, jun. (*J. Met.*, 1960, **12**, Feb., 124). **The general case, titanium** W. W. Minkler (124–126). **Case for steel** R. B. Gunia (126–127). **Airframes titanium** E. A. Simkovich (128–129). **Steel in airframes** E. A. Green (129–131). **Jet engines—titanium** G. S. Wile (132–133). **Steel in jet engines** R. E. Patsfall (134–136). **Pressure vessels—titanium** A. Hurlich (136–138). **Steel in pressure vessels** J. F. Baisch (138–139). **Solid-fuel motor cases, titanium** J. Sohn (139–141). **Steel in solid-fuel motor cases** D. K. Hanink (142–143). **Applications, chemical and marine, titanium** G. R. Prescott (143–146). **Metals in chemical and marine applications** F. W. Fink (146–148).

Design information on 5 Cr alloy steels for aircraft and missiles F. R. Morral, R. J. Favor, and W. P. Achbach (*DMIC Report* 116, *PB* 151072, 1959, Aug. 28, pp.42) H-11 type steels have high strength-density ratio at ~1000°F, sufficient hardenability for air quenching and a tempering temp. of 1000–1100°F. They lack corrosion resistance, are susceptible to notch effects, may have low transverse ductility, and the mechanical properties are sensitive to carburization and decarburization. Data from published reports and papers are collected.

Free machining stainless upgrades missile parts Perfecting Service Co. (*Steel*, 1960, **146**, Jan. 25, 69) Process advantages from making an important threaded missile component from free machining stainless steel instead of Cr–Mo steel are described.—D.L.C.P.

Demand for missile materials spurs metals research W. S. Lyman (*Iron Age*, 1959, **184**, Nov. 19, 124–126) The role of missile development in stimulating metallurgical research is discussed. Problems posed include extreme temp., rapid fluctuations in load, and the effects of reactive or corrosive fuels.—A.G.

Development of iron-aluminium base alloys for gas cooled reactor components. Second quarterly progress report J. Mueller (*MND–DB*–2522, 1958, June, pp.86; from *US Res. Rep.*, 1959, **32**, Aug. 14, 278; no abstract).

Development of iron-aluminium base alloy for gas cooled reactor components: quarterly progress report No.4 for Sept. through Nov. 1958 F. Tate (*MND–DB*–2524, 1958, Dec., pp.49; from *US Res. Rep.*, 1959, **32**, Aug. 14, 278; no abstract).

Stainless steel—uranium dioxide fuel components for army package power reactor J. E. Cunningham and R. J. Beaver (*Nuclear Met.*, 1958, **5**, 29–40) Fabrication is described.

HISTORICAL

Abraham Darby and the coke-smelting process M. W. Flinn (*Economica*, 1959, Feb., 54–59, reprint) Arguments for a later date than 1709 are presented.

Coalbrookdale 1709 to 1959 A. Raistrick (*Found. Trade J.*, 1959, **107**, Oct. 1, 239–247) An illustrated historical survey.

British ironmakers and ironfounders abroad M. Schofield (*Found. Trade J.*, 1959, **107**, Dec. 10, 589–590) Brief biographies of John Hughes of Stalino and John Wilkinson of Nantes and Le Creusot.

Castings for the Colonies—America's first iron works (*Mod. Cast.*, 1960, **37**, Jan., 48–49) The history of the Saugus Ironworks, which has been restored, is outlined.

The casting of grey iron, before the introduction of the new materials and methods of manufacture. II D. Iturrioz (*Met. y Electr.*, 1960, **24**, March, 75–84) A comparison of past with present methods in: (1) Coke furnaces, blast-furnaces, sintering furnaces, foundries, homogenizing kilns, reheating furnaces, rolling plant, competition with other materials and methods and competition between grey iron and other materials and methods. A general review.

The beginning of sheet rolling in Saxony (1823–1827) E. Matthes (*Stahl Eisen*, 1960, **80**, March 31, 437–441) Historical.—T.G.

ECONOMICS AND STATISTICS

The economic topography of iron and steel in Europe E. Reiffer (*Nachr. Eisen–Bibliothek Georg Fischer AG*, 1959, (18), Dec., 69–73) A summary of a lecture by H. Boesch to the second Eisen-Bibliothek meeting in Nov. 1959 dealing with the economic trends now emerging clearly in the iron and steel industry.—T.G.

The iron balance in the national economy L. L. Zusman and A. A. Brodov (*Stal'*, 1960, (2), 160–164) Mainly statistical, and referring largely to non-recoverable losses in smelting, use, and recovery, including corrosion losses.

The industrial situation. Iron and steel production in France in the course of 1959 (*Usine Nouv.*, 1960, **16**, Jan., 14 11–12) An analytical review of production figures, steel prices, the market for special steels in 1959, and the prospects for 1960.—S.H.-S.

The future of the Soviet iron and steel industry P. I. Korobov (*Stal'*, 1960, (1), 1–4) A discussion of the progress of the seven-year plan during 1959, with statistics.

Soviet iron and steel. The drawbacks of wide open spaces (*Times Rev. Ind.*, 1959, **13,** Dec., 72–73) With sketch map.

The mineral production of the Indian Union during 1949 M. S. Krishnan (*Records Geol. Survey India*, 1954, **83,** (3–4), 521–702). **Ilmenite** (573–574). **Iron-ore, pig-iron and steel** (574–578). **Consumption of iron and steel materials** (579) Statistical. **Manganese-ore** (585–589).

Japan's iron and steel industry 1958. Summary (*Japan's iron and steel industry* 1959, 89–98) Statistics and prices. **Exports of iron and steel in 1958** Japan Iron and Steel Exporters' Association (99–107) Statistical. **Galvanized iron sheets 1958 production and exports** (124–125). **Wire products 1958 production and exports** (126–128). **Special steel industry 1958 summary** (130–133). **Ferro-alloy industry** (135). **Casting and forging industries** (136–139). **Statistics** (Part IV, 193–224). **Directory of iron and steel makers** (Part V, 259–276).

MISCELLANEOUS

Why is a metallurgist? L. M. Hogan (*Austral. Eng.*, 1959, **52,** Nov., 49–55, 69) A review of the work of metallurgists in various capacities, especially in research, and its application in industry.

The role of the industrial engineer in the steel industry J. E. Angle (*Iron Steel Eng.*, 1959, **36,** Aug., 63–66) The development of industrial engineering, its relation to the organization, and the function it should be assigned if maximum benefits are to be derived are discussed.

Cost estimating—metal working plants A. J. Mosso (*Iron Steel Eng.*, 1959, **36,** Sept., 87–99) It is shown and illustrated by means of a detailed example how a preliminary cost estimate can be made which is calculated by the metal working engineer and is sufficiently accurate for appropriation requests.

Electrowinning tungsten and associated molybdenum from scheelite J. B. Zadra and J. M. Gomes (*US Bur. Mines, Rept. Invest.*, 554, 1959, pp.23).

The crystal structure of iron (II) chloride tetrahydrate B. R. Penfold and J. A. Grigor (*Acta Cryst.*, 1959, **12,** Nov., 850–854).

Crystal lattice and space group of anhydrous ferrous sulphate J. Coing-Boyat (*Acta Cryst.*, 1959, **12,** Nov., 939) [In French].

The distillation of mixed chlorides obtained by the chlorination of ferro-niobium J. S. Broadley and B. R. Steele (*RDB(C)TN*–109, 1955, Jan., pp.8; from *Nucl. Sci. Abs.*, 1959, **13,** May 31, 1167) Pure $NbCl_5$ could be obtained by this process.

Arsenides of the transition metals. III. A note on the higher arsenides of iron, cobalt and nickel R. D. Heyding and L. D. Calvert (*Canad. J. Chem.*, 1960, **38,** Feb., 313–316) Crystal structures are given from powder diffraction studies.

Microstructure of silica in the presence of iron oxide L. H. van Vlack (*J. Amer. Ceram. Soc.*, 1960, **43,** March, 140–145) Quartz and haematite were heated together up to 2800°F and structures are shown.

Electronic data processing W. K. Smith (*Iron Steel Eng.*, 1959, **36,** Sept., 160–166) The mechanism of the electronic computer is described in outline, and its applications and other systems for data processing are described, with particular reference to methods in use at the Steel Division of the Ford Motor Company at Michigan.

Present status and trends in data processing and programming in the steel industry R. M. Sills and G. E. Terwnnger (*Iron Steel Eng.*, 1959, **36,** Aug., 72–78) The authors describe possible methods of data processing as applied to the control of physical measurements, in relation to systems in use and to trends of development. Applications to the sintering plant, blast-furnace, open-hearth furnace, hot mills, and the production and processing of strip are outlined.

Solving operating problems with electronic computers W. O. Blattner (*Blast Furn. Steel Plant*, 1960, **48,** Feb., 183–187) A review of various techniques and types of problems for which electronic computers are used.

Pressure-drop measurements through beds of granular iron, lathe turnings, steel wool and parallel-plate assemblies R. M. Jimeson and W. A. Decker (*US Bur. Mines Rep. Invest.*, 1959, (5541), pp.51) The data are intended for Fischer–Tropsch process studies.

Consider automation growth key to plant modernization A. J. McLaren (*Iron Age*, 1959, **184,** Oct. 8, 105–107) The need for metal workers to investigate the possibilities of automation is stressed. Standard unit components are giving much greater flexibility to automatic operations and thereby making them profitable to smaller companies.—A.G.

The attitude of the workmen in the iron and steel industry against innovations in running the workshop E. A. Jüres (**Stahl Eisen*, 1960, **80,** March 31, 429–437) This is a contribution to the European Productivity Centre. To sum up this rather philosophical account it can be said that the workmen—in the majority of cases—do not mind changes, i.e. modernization, partly automation, and the like, being introduced in their workshop, as long as they are informed beforehand and given the proper training for the new mode of operation.—T.G.

BOOK NOTICES

AITCHISON, L. '*A History of Metals*'. Two volumes. 11 × 8½ in., pp.xxi + 303, xvii + 344 Illustrated. 1960. London: Macdonald & Evans Ltd. (Price, two vols. £8 8*s*.)

Written by a very competent metallurgist with an unusually wide knowledge of his field, this work is far more than a mere history of metallurgy itself. Linked as it is with a sound, though never verbose, discussion of the historical background against which metallurgy developed, these volumes become a history of the material evolution of human society in general.

Starting from the neolithic age, the author traverses the whole sequence of development until his two last illustrations are a billet of plutonium and a section through a sample of neptunium respectively. With such an immense field, extending over some six or seven thousand years, there is a real danger of the treatment degenerating into superficiality. That this is far from being the case no one to whom the author is known will need to be reassured. Not only the metallurgist but also the historian and archeologist will find real meat into which to get his teeth. But—and it is a big but—the presentation is so happy, and the large number of illustrations so well chosen—and incidentally reproduced—that it may be read with pleasure by any intelligent reader; freedom from technicalities is one of its outstanding virtues.

In volume I which carries the story down to about 1000 AD, iron plays a somewhat subsidiary role; in the second this state of affairs is reversed. First wrought- and then cast-iron and later steel take the centre of the stage but supported by a strong cast of secondary (non-ferrous) parts. 'When the 13th Century opened, iron—in most parts of Europe and certainly in England—was still being extracted from its ores by almost exactly the methods that had been used three millenia earlier'. Further, 'there are good reasons for stating that metallurgy was mediaeval until the 18th Century', while in Figure 249 is reproduced for the first time a letter from Sorby to Sir Robert Hadfield. It is a remarkable feat to have welded all this together to form a coherent story which is so well written, well illustrated, excellently produced, and (thank goodness) well indexed.—F. C. THOMPSON.

ALÈGRE, R. *et al.* '*Etudes de Pyrométrie Pratique*'. Préface du G. Ribaud. Collection de l'A.N.R.T. 9¼ × 6 in., pp.231. Illustrated. 1959. Paris: Eyrolles, Gautier-Villars. (Price 3300 Fr.)

This book comprises four essays by a panel of eight authors on particular branches of temperature measurement where recent advances have been made. The first part is a brief survey of the different techniques of temperature measurement confined to headings and very short descriptive notes. The second part deals with the use of thermistors as a special case of resistance thermometry. The circuitry considered is equally applicable to platinum resistance devices. Part III deals with the general application of thermocouples illustrated for the particular case of platinum *v.* platinum/rhodium combinations. There is a very good description of the possible sources of error and the means by which their effect may be reduced. Reference tables of the outputs from various combinations of couples are given but more information on the advantages of different combinations would have been better, e.g. the insensitivity of the 20% and 40% rhodium combination to cold junction temperature change is not commented upon. The longest and best part of the book is devoted to optical pyrometers and in particular to flame temperature measurement using photocell receivers. A valuable description of the different techniques needed for various types of flame and details of pyrometers is given. A section of this paper is devoted to suction pyrometers which is certainly relevant and useful, but inconsistent with the title of the paper.—J. R. PATTISON.

BAILEY, A. R. '*A Text-Book of Metallurgy*', 2nd ed. 8½ × 5½ in., pp.x + 561. Illustrated. 1960. London: Macmillan & Co. Ltd. (Price £1 10*s*.)

In this revised edition the text has been modified in parts, and in some cases rewritten. In general the book retains its original composition. The first 176 pages are based on various aspects of physical metallurgy, the next third is devoted to extraction and refining processes of ferrous and non-ferrous metals, and the rest of the book is made up of chapters on mechanical testing, working, joining, powder metallurgy, and pyrometry. Many important binary diagrams appear in one of four appendices. Another deals at some length with methods of writing metallurgical reports. At the end of the book there is a large collection of questions selected from past examinations over a range from ordinary national certificate level to final honours of several Universities. The wide field that has been covered has not permitted of anything more than an introduction but the list of additional reading at the end of each chapter should be of great benefit. The author has been very ambitious in his endeavour to include so much in a book of about 570 pages. He has definitely provided the student with a good general appreciation of the principles and practice of metallurgy and the book can be highly recommended.

A. J. MACDOUGALL.

CHALMERS, B. '*Physical Metallurgy*'. (Wiley Series on the Science and Technology of Materials). 9 × 6 in., pp.xviii + 468. Illustrated. 1959, New York; London: John Wiley & Sons. Inc.; Chapman & Hall Ltd. (Price £5.)

During the last few decades physical metallurgy of the descriptive type has given place progressively to a more systematic and quantitative approach with ever greater emphasis on underlying scientific principles. New and important experimental techniques have played a vital part in making these changes possible and in providing the evi-

dence to substantiate new ideas and theories. In fact, techniques play such a vital part in modern physical metallurgy that too much importance is sometimes attached to techniques *per se*, and too little to the information they provide about metals, their properties and behaviour. In the book under review, Professor Chalmers has deliberately gone to the other extreme. He has written an introductory text-book on physical metallurgy in which no attempt is made to describe experimental techniques. Instead, the results of these techniques are discussed as and where necessary to build up a coherent and logically developed picture of physical metallurgy, written from the modern scientific point of view, and yet introducing much of what is valuable in the classical approach. Starting with a discussion of the atomic structure of the metals he considers metallurgical phenomena from the point of view of the atomic processes involved, and correlates the quantitative and the descriptive viewpoints wherever possible.

Some might criticise the treatment as being somewhat superficial, but this would be to ignore the point of view from which the book has been written. As he makes clear in his preface, the author prefers students to be introduced to scientific physical metallurgy after they have already reached a fairly advanced stage in the basic physical sciences. Then too, the student is expected to read elsewhere both to learn about experimental techniques and to acquire the quantitative detail that the author has deliberately omitted in order that the essential features shall stand out the more clearly. Students who take this advice will acquire a good knowledge of physical metallurgy and will find this book an excellent guide to the subject, and one that will make them appreciate how the different pieces of the jigsaw of physical metallurgy fit together. Those who do not do the supplementary reading, but expect the book to stand on its own, may well take the view that a fuller treatment of some of the more complex topics is desirable; they should read or re-read the preface. More experienced metallurgists who have not kept pace with modern developments will find this book of value in giving them, in short compass, an authoritative bird's-eye view of modern physical metallurgy, and possibly as a handy reference book to which they can turn for clear and concise explanations of terms such as 'thermal spikes' or 'sessile dislocations'. Each chapter concludes with advice on further reading and an appendix contains selected problems to test the student's knowledge and to encourage his studies.

Altogether this represents something new in metallurgical text-books, and it should help many to see modern physical metallurgy in perspective. It is unfortunate that the price (£5) is so high that many of those who would benefit most from it will feel that it is beyond their purse.—A. G. Quarrell.

Elliot, J. F. and Gleiser, M. '*Thermochemistry for Steelmaking*'. Vol. I. 11 × 8½ in., pp.viii + 296. Illustrated. 1960, Reading, Mass. and London: Addison-Wesley Publishing Company Inc. (Price $10.50.)

This is the first in a series of volumes being prepared in the Department of Metallurgy of the Massachusetts Institute of Technology, and sponsored by the American Iron and Steel Institute in order to collect, consolidate, and summarize in one place all the data pertinent to the physical chemistry of steelmaking. This first volume is intended as the foundation on which the material for subsequent volumes will be developed. It is presented in four sections, dealing in turn with physical properties of selected elements, thermochemical properties of selected elements, standard heats and free energies of formation of selected compounds, and vapour pressures of selected elements and compounds. No data is given on the enthalpy increments, entropy increments, and heat capacities of compounds because the authors are awaiting a new publication by Dr K. K. Kelly. The data is presented in clear tabular or graphical form with adequate references to the main sources of information. Temperatures quoted in the tables are very rightly given in °Kelvin but graphs are drawn with a Fahrenheit scale over which has been superimposed a Centigrade scale. It is promised that a table of temperature equivalents will appear in a later volume. A comprehensive index enables the reader to find any particular information with ease, though the use of the symbol Ar for Argon may temporarily confuse some readers. One important outcome of a collection of data of this sort is that it brings to light the gaps in the information. This volume makes clear that there is a serious lack of data for nitrides, carbides, and sulphides, and one hopes that this fact will stimulate further research. The authors have made a good start to their work and it is to be hoped that subsequent volumes will not be too long delayed.—E. Ineson.

Elsas, M. (Editor). '*Iron in the Making. Dowlais Iron Company Letters 1782–1760*'. 9¾ × 6 in., pp.xix + 247. Illustrated. 1960, Cardiff: Glamorgan County Council and Guest Keen Iron & Steel Co. Ltd.

Few industrial concerns can carry their history back for 200 years; and none, it is safe to say, has marked the opening of its third century more fittingly than the Guest Keen Iron & Steel Company. The directors have shown large-mindedness and public spirit by depositing their massive records, for the use of students, in the Glamorgan County Record Office; in this volume they offer to a wider public a sample of some 650 items drawn from more than 500,000 letters in the collection at Cardiff. The letters tell the story of the relations, competitive and co-operative, between the Dowlais partners and other pioneer firms in South Wales; of the fluctuations caused by the ebb and flow of demand from the Navy, and later from the railways, for Dowlais iron; of developments in techniques and transport; and of the course of local politics. A vivid impression is obtained of the difficulties of English employers who, in the early years at least, understood neither the language nor the temperament of the majority of those who worked for them; if there is much in the letters about strikes and riots, there is much also about successful efforts to create industrial peace and goodwill. The people one learns most of are the able and intrepid proprietors, William Taitt and Sir Josiah John Guest; but one gets glimpses also of eminent technologists and scientists, such as Peter Onions, David Mushet, Robert Stephenson, I. K. Brunel, and Michael Faraday. It was a wise decision to treat the reader as an adult and leave the facts, whether creditable or discreditable, to speak for themselves. The work of selecting and editing the letters has been done with outstanding skill by the Glamorgan County Archivist, Madeleine Elsas, whose brief notes, family pedigrees, and biographical index give the reader all he needs to appreciate the background of the correspondence. The book makes a substantial contribution to scholarship: it should find a place in the libraries of all concerned with the history of iron and steel and the industrial revolution in Britain.

—T. S. Ashton.

Leymonie, C. '*Les Traceurs Radioactifs en Métallurgie Physique*'. Préface de P. Lacombe. 8½ × 5½ in., pp.xvi + 221. Illustrated. 1960, Paris: Dunod (Price 32 NF.)

In recent years radiotracer techniques have contributed in many fields of physical metallurgy, and the present book, aimed primarily at the metallurgist and solid state physicist, gives a comprehensive and up-to-date review of the application of these techniques to metallurgical problems.

The first chapter, headed 'Généralités' gives a brief but adequate account of the relevant properties of atoms, isotopes, isomers, etc. It includes a particularly useful, section devoted to the preparation, selection and use of isotopes in radiotracer work. The remainder of the book is divided into four chapters on specific physical metallurgical topics to which tracer techniques have made major contributions. Diffusion, of course, merits the largest chapter and 80 pages of the book are devoted to this subject. After a short introduction dealing with diffusion theory, the implications of tracer work on such aspects as self diffusion, diffusion at infinite dilution, effect of strain on diffusion, and diffusion in alloys and liquids are discussed. Subsequent chapters describe respectively segregation during solidification or in the solid state, surface reactions, and equilibria between a vapour and solid phase or in slags.

The book is recommended as a concise and valuable source of information on the present position of radiotracer applications in physical metallurgy. Detailed descriptions of experimental techniques or arrangements are not included, but to each chapter a profusion of references is appended, e.g. 178 on diffusion, so that original may be readily followed up if required.—A. G. Knapton.

Metallurgical Society [of AIME] Conferences. Vol. I edited by T. E. Dancy and E. L. Robinson: '*Flat Rolled Products: Rolling and Treatment*'. pp.xix + 128. (Price £1 10*s*.). Vol.2 edited by W. R. Clough: '*Reactive Metals*', pp.xiv + 610. (Price £5 13*s*.). Vol. 3 edited by R. W. Lindsay: '*Quality Requirements of Super-Duty Steels*', pp.x + 309. (Price £3 4*s*.). Vol. 4 edited by R. N. Rhodin: '*Physical Metallurgy of Stress Corrosion Fracture*', pp.xiii + 394. (Price £4 18*s*.). 9 × 6 in. Illustrated. 1959, New York; London: Interscience Publishers.

Each volume of this new series (individual papers are being abstracted in the *JISI*) constitutes the proceedings of a technical conference sponsored by the Metallurgical Society of AIME or one of its technical committees. The idea is an excellent one. So many meetings of great importance to the steel industry are now being held throughout the world that it has become a serious problem to keep track of them. Moreover, the availability of printed reports afterwards is sometimes severely restricted. This publication can be welcomed on more than one score, for it promises also to provide first class contributions to the technical literature.

Volume 1 reports the proceedings of a conference on the rolling and treatment of flat rolled products which was held at Chicago in January 1959. It contains six papers each with discussion and it gives also the report of a panel discussion on surface texture. Volume 2 reports a conference held at Buffalo, New York, in May 1958, on reactive metals, here defined as Ti, Zr, Hf, Th, V, Nb, Ta, Mo, W, U, Re. A total of 37 papers is presented, among them several describing processes and procedures which are common to several of the reactive metals: for example, consumable electron beam meeting, powder metallurgy, and the micro-gravimetric absorption method.

Volume 3 is a report of a conference held at Pittsburgh in May 1958, on the properties ('quality requirements') of 'Super-Duty' steels, a term which appears to include any steel having a non-traditional application: it was decided at the outset that the term embraced steels which have to resist unusual and severe environments, or which have chemical, mechanical, and physical properties that are much superior to those of conventional carbon steels. With such a vast and still largely uncharted field before it, this conference covered much useful ground and the 15 papers accompanied by discussion are of a very high standard. Volume 4 reports a symposium held at Pittsburgh in April 1959 on the physical metallurgy of stress corrosion fracture. There are 16 main contributions with discussion and, as in the other volumes in this excellent series, the editorial work has been done most efficiently.

M. L. Pearl.

Morgan, J. H. '*Cathodic Protection. Its theory and practice in the prevention of corrosion*'. 10 × 6½ in., pp.xviii + 325. Illustrated. 1959, London: Leonard Hill (Books) Ltd. (Price £2 17*s*. 6*d*.).

Wiederholt, W. and Kaesche, H. '*Korrosion* II *Kathodischer Korrosions-*

schutz'. Proceedings of a symposium in Frankfurt-Main, 11–12 February 1958. 8½×5½ in., pp.xvi+229. Illustrated. 1959, Weinheim/Bergstr.: Verlag Chemie. (Price DM.30.–).

APPLEGATE, L. M. '*Cathodic Protection*'. 9½×6¼ in., pp.ix+229. Illustrated. 1960, New York; Toronto; London: McGraw-Hill Book Company Inc. (Price £3 10*s*. 0*d*.).

Of the above three publications that by Morgan is most informative and appears to be the first English text-book on cathodic protection. The theoretical simplicity of the principle on which this process operates offers the possibility of complete freedom from corrosion as a practical target. All other anti-corrosive measures seem by contrast to be rear-guard actions saving as much as possible for as long as possible.

Morgan's text-book is well planned and its lists of figures and plates immediately after the contents list will save students time. The 11 chapters commence with one on the historical and theoretical aspects but quickly get down to practical cathodic protection parameters. The line drawings are clear and the photographs excellently reproduced. The book explains in a practical manner the techniques employed in this method of preventing corrosion. Chapters 2 to 5 treat cathodic protection as a technique; chapters 6 to 9 consider methods of achieving protection on a wide variety of structures; chapters 10 and 11 deal with instrumentation and economics. The bibliographies at the end of each chapter should help new students of the subject but it seems surprising to find no reference to the work of the 'Joint Committee for the Co-ordination of the Cathodic Protection of Buried Structures' dated 1957.

The German book is smaller; a handy, neatly printed report of a two-day symposium held in Frankfurt-Main during February 1958 and organized by the Committee for Corrosion and Corrosion-Prevention of the Deutschen Gesellschaft für Metallkunde. This was the fourth of a series of yearly symposia in Germany and corresponded closely with that on the same subject held in 1953 by the S.C.I. Corrosion Group in London. The chapters are the work of separate German authors. H. Steinrath commences with an historical introduction; H. Kaesche deals with theoretical principles; W. Rausch recommends the use of magnesium electrodes for protection against corrosion caused by fresh waters; J. Lepper considers the use of sacrificial electrodes in preventing sea water corrosion; H. Klas says there has been an unexpectedly rapid development of cathodic protection for steel pipes and other constructions buried in the ground. So the book proceeds with further sections by H. Riedel, W. von Baeckmann, K. Thalhofer, and G. Schikorr; concluding with a bibliography by W. Wiederholt and H. Kaesche, 34 pages long and divided into 16 sections.

The American book is the work of a professional engineer, a retired Lieutenant Colonel with experience of an inspection assignment in which he observed corrosion problems from Maine to Okinawa. It sets out to be a guide to engineers who are advised to do-it-yourself. So the author gives practical how-to-do-it examples for the application of the methods of cathodic protection to buried pipelines and cables, aviation fuel tanks, lock and dam gates, laid-up ships, and so on. A refreshing, if not unduly scientific attitude pervades the book which strongly advises engineers to get on and do something themselves about preventing corrosion by the techniques of cathodic protection.

These techniques of preventing the unwanted corrosion of metals are being studied and employed in many parts of the world. The literature of the subject has grown rapidly in the form of papers and brochures. The simultaneous appearance of these three books underlines the growing importance of the subject.

—T. HENRY TURNER.

NEWTON, J. '*Extractive Metallurgy*'. 9×6 in., pp.vii+532. Illustrated. 1959, New York; London: John Wiley & Sons Inc.; Chapman & Hall Ltd. (Price £3 18*s*.).

This book is essentially a revised version of a part of the 1947 edition of Newton's 'Introduction to Metallurgy'. An attempt has been made to discuss principles rather than give detailed descriptions of plant and practice. In a limited way the author has been successful but his treatment is often fragmentary. The chapter on ore dressing is reasonably well done at an elementary level but the chapter on chemical principles is disappointing. A student reading metallurgy should already be familiar with simple chemical equations and the chapter could have been improved by giving more space to the physical chemistry of process metallurgy. Fifty-seven pages of physical metallurgy seems rather a lot to incorporate into a book on extraction. Phase diagrams have their application but a discussion of topics such as Miller indices, metallic bonding, or the martensite reaction seems to be pointless. The 'Extractive' portion of the book is contained in three chapters which also include refining processes such as steelmaking by convertor, open-hearth, electric arc, and induction furnaces.

The book is printed in clear type, well illustrated, and carefully chosen references are given at the end of each chapter.

As a general introduction to 'Extractive Metallurgy' the book probably fulfills its function in a superficial way but its 532 pages could have been used to better advantage by the omission of many extraneous topics.—A. J. MACDOUGALL.

PARR, N. L. '*Zone Refining and Allied Techniques*'. With a foreword by E. J. Vaughan. 8½×5½ in., pp. xvi+184. Illustrated. 1960, London: George Newnes Ltd. (Price £2.).

The process of 'zone refining', based on simple physical principles and applicable with the minimum of manipulation, arrived at a time when the important effect of trace impurities on specific properties of materials became fully appreciated. Developed by W. G. Pfann in the first place for the purification of germanium in semi-conductors, the technique has been eagerly investigated and adopted by many workers in the purification of elements, and of both organic and inorganic compounds. In this book the author has collected the large amount of information which has accumulated since the original discovery in 1951. Following the stated intention to deal with the practical aspects of the technique, only a brief account of the theoretical principles is given, and the main section (approximately 150 pages) comprises a survey of practical information in recent advances in laboratory methods and equipment. Consideration is given to design of equipment and many examples of apparatus are shown, as well as a description of manipulation methods both before and after treatment by zone-refining. The technique of crystal growing, which is important in the production of semi-conductors, is described. In the final two chapters, an account is given of the examination and application of high-purity substances. The text is supported with numerous line drawings of excellent quality which illustrate the design and operation of apparatus for dealing with materials ranging from the metals to organic compounds. On the grounds of similarity, perhaps a few of the drawings might have been omitted. There are also 24 photographic plates showing apparatus and products of zone-refining techniques, all of which are informative with the exception of plate IV which merely shows a typical metal cabinet containing electronic equipment. For those who wish to pursue the subject further, adequate references are given at the end of each chapter, and there is an efficient index. The type is in the clear style uniform with this series of practical science books. This book may be read with advantage by those wishing to acquaint themselves with the rapid developments of a new and important laboratory technique, and also by those workers who, requiring materials in a high state of purity, may find a ready means in the employment of the zone-refining method.—R. HANDLEY.

REINER, M. '*Deformation, Strain and Flow*'. 2nd ed. 9½×6½ in., pp.xvi+347. Illustrated. 1960, London: H. K. Lewis & Co. Ltd. (Price £3 3*s*.).

The deformation and flow laws for a wide variety of materials and liquids are discussed mainly from a macroscopic, phenomenological standpoint. Three basic ideal types are first defined which are called the Hooke elastic solid, the Newtonian viscous liquid, and the St Venant plastic solid. After emphasizing that no real substance can be so narrowly defined, the author goes on to consider many combinations of these basic types (e.g. visco-elastic, Maxwell liquid, Bingham solid, etc.), to develop the basic equations governing their behaviour, and to apply these to various phenomena and problems. Certain secondary phenomena are also treated, such as the Weissenberg effect. The book should be useful for anyone who wants an overall view of a wide range of mechanical behaviour occurring in liquids and solids. However, so much ground is covered that the treatment of any one topic is bound to be somewhat unsatisfactory for specialists. Nor is it always adequate as an elementary introduction; for example no mention is made of stress strain-increment relations in the chapter on plasticity. The text is based on lecture notes. The resulting discursive style, the use of the first person singular (for which the author apologises), and the constant intrusion of the personality and research career of the author, make this an unusually personal book, which may or may not please the reader.—A. P. GREEN.

SCHULENBURG, A. '*Handbuch der Schmelz- und Legierungspraxis in der Metallgiesserei*'. 2. Aufl. 8×5½ in., pp.371. Illustrated. 1959, Berlin: Fachverlag Schiele und Sihön GmbH. (Price DM.28.–).

This practical handbook of 350 pages first published in 1952, has enjoyed a measure of popularity in Germany such that it has now been reprinted. The 1959 edition has been modernized and revised where necessary, in particular the sections on melting and alloy composition. This book is essentially a practical man's reference book and does not purport to be anything else. It is written primarily for the apprentice or skilled foundryman and for the ever-growing group of semi-skilled foundry engineers.

It is written in German for the German foundry craftsman and therefore metal compositions are those given in German standard specifications and all the literature references are to German authors. Therefore, this book must be of limited value in Britain where the reader who would most benefit by it can hardly be expected to know sufficient of the language to read it. Furthermore, the few illustrations are to German equipment exclusively and the diagrams are not of a sufficiently high standard to be self-explanatory. If, however, the barrier of language and customs can be overcome, the book has much to recommend it and its essentially practical bias might well be a model for a similar book here. No useless historical matter is included, no references to obsolete methods and processes; theory has been kept out except where absolutely essential to the proper understanding of the text. It is a book for the harassed foundryman, faced with practical difficulties, and a good text-book for those teachers of foundry craft with an insufficiently practical background. It covers the entire metals casting industry, iron, steel, non-ferrous, light alloys, special purpose alloys, and newer metals.—R. V. RILEY.

VICKERY, B. C. '*Classification and Indexing in Science*'. 2nd ed. 8½×5½ in., pp.xix+235. 1959, London: Butterworths Scientific Publications. (Price £1 10*s*.).

It is not very hard to sort cards according to a certain system and file them in an index. The difficulty is in making sure that they can all be found again when they are wanted, and Mr Vickery authoritatively, describes the present state of this art of

'information retrieval'. The simplest index for the untrained user is the alphabetical one but it is dependant on numerous cross-references. Classified schemes, potentially more efficient, break subjects down by 'facet analysis' into their components and arrange these in a helpful order; a notation is then applied which will maintain this order and allow for new subjects; to complete the system an alphabetical index is necessary as a key to aspects which the notation does not reveal. Finally Mr Vickery deals with mechanical techniques, emphasizing that (i) they are only as good as the classification schemes on which they are based; (ii) information machines can do nothing that humans cannot, and it is not yet certain whether they are more efficient.

Detailed examples are given throughout, and the author defines his terms so that this book could be used as an introduction to the subject; but preliminary readings in the end-of-chapter bibliographies would be helpful to the uninitiated.—M. WRIGHT.

WESTWOOD, W. and MAYER, A. '*Chemical Analysis of Cast Iron and Foundry Material*'. 2nd. ed. revised by W. E. Clarke and H. Green. Foreword by E. Gregory. $9\frac{1}{4}\times6\frac{1}{4}$ in., pp.586. Illustrated. 1960, London: George Allen and Unwin Ltd. (Price £3 3*s*.).

Well-thumbed first editions of 'Westwood and Mayer' have survived nine years of constant consultation in routine foundry laboratories with surprisingly little damage. The new edition is similarly strongly bound and contains 20 more pages of text printed on even better paper. The original authors having undertaken new responsibilities, the revision was entrusted to Mr W. E. Clark F.R.I.C. and Mr H. Green, who have followed them into similar posts in the laboratories of the British Cast Iron Research Association. The book opens with a general account of chemical practice and then deals in turn with the sampling and analysis of iron, ferro-alloys, slags, silica, coal, coke, foundry binders, etc. Finally an appendix of useful tables and works of reference are included. The form of the original edition is retained but the analytical methods have been modified or replaced in the light of experience gained not only at the British Cast Iron Research Association but by dozens of analysts engaged in work in the same or similar fields. More and more trace elements are now known to influence the behaviour of cast iron both in the foundry and in use; therefore, their determination has been given additional attention in this edition. The new edition costs only 63*s*., and is essential to all practising metallurgical chemists.

—R. V. RILEY.

NEW PUBLICATIONS

These titles are also available on index cards in advance of publication in the Journal, *each title being classified under the Universal Decimal system (UDC). Details and subscription rates can be obtained from the Secretary, The Iron and Steel Institute.*

AITCHISON, L. '*A History of Metals*'. Two volumes. $11\times8\frac{1}{2}$ in., pp.xxi+303; xvii+305–647. Illustrated. London, 1960: Macdonald Evans Ltd. (Price, two vols. £8 8*s*.)

AKADEMIYA NAUK SSSR. Institut Kristallografii. [Academy of Sciences, USSR. Institute of Crystallography.] '*Growth of Crystals*'. Reports at the First Conference on Crystal Growth, 5–10 March 1956. In English translation [original '*Rost Kristallov*', Moscow, 1957.] $10\frac{3}{4}\times8\frac{1}{4}$ in., pp.294. Illustrated. New York, 1958: Consultants Bureau Inc. (Price £6: $15.00.)

AKADEMIYA NAUK SSSR. Institut Machinovedeniya. [Academy of Sciences, USSR. Institute for the study of machines.] '*Voprosy prochnosti materialov i konstruktsii*'. ['*Questions of the strength of materials and structure*'.] In Russian. $10\frac{1}{2}\times6\frac{1}{2}$ in., pp.401. Illustrated. Moscow, 1959: The Academy (Price 20r. 90k.)

AKADEMIYA NAUK SSSR. Sibirskoe Otdelenie. Institut geologii i geofizik. [Academy of Sciences, USSR. Siberian division. Institute of Geology and Geophysics.] A. N. Vorsin, E. F. Doil'nitsyn, A. I. Trubetskoi, and M. Ya. Shcherbakova. '*Radiochastotnyi mass-spektrometr (teoriya, raschet i konstruirovanie*'. ['*The radio frequency mass-spectrometer (theory, calculation and construction*'.] In Russian. $8\frac{1}{4}\times5\frac{3}{4}$ in., pp.76. Illustrated. Moscow, 1959: The Academy. (Price 2r. 70k.)

AKADEMIYA NAUK SSSR. Sovet po izucheniya proizvoditel'nykh zil. Ministerstvo Geologii i Okhrany Nedr SSSR. Chitinskoe geologicheskoe upravlenie. [Academy of Sciences, USSR. Council for the study of production resources. Ministry of Geology and preservation of Mineral Resources, USSR. Chita geological authority.] '*Poleznye iskopaemye Chitinskoi oblasti (Chernye metally i nemtallicheskie poteznye iskopalmy)*'. ['*Mineral resources of the Chita region (ferrous metals and non-metallic minerals)*'.] In Russian. $10\times6\frac{1}{4}$ in., pp.143. Illustrated. Moscow, 1959: The Academy. (Price 7r. 85k.)

AMERICAN SOCIETY FOR TESTING MATERIALS. '*ASTM Standards on Metallic-coated iron and steel Products*'. Sponsored by ASTM Committee A–5 on Corrosion of iron and steel. Zinc-coated sheets, hardware, wire, strands, fencing, and pipe, aluminium-coated products, Terne-alloy-coated sheets. 9×6 in., pp.viii+164. Illustrated. Philadelphia. 1959: The Society. (Price, Members $2.80; Non-members, $3.50.)

AMERICAN SOCIETY FOR TESTING MATERIALS. '*ASTM Standards on methods of atmospheric sampling and analysis*'. Prepared by ASTM Committee D–22 on methods of atmospheric sampling and analysis. Methods of testing. Recommended practices. Definitions. 9×6 in., pp.vii+102. Illustrated. Philadelphia, 1959: The Society (Price, Members $1.80; Non-members $2.25.)

AMERICAN SOCIETY FOR TESTING MATERIALS. '*Symposium on Electron Metallography*'. Presented at the Sixty-second Annual Meeting, Atlantic City, N.J., June 23 1959. S.T.P. No.262. 9×6 in., pp.128. Illustrated. Philadelphia, 1960: The Society. (Price, Members $3.40; Non-members $4.25.)

AMERICAN SOCIETY FOR TESTING MATERIALS. '*1958 References on Fatigue*'. Sponsored by ASTM Committee E–9 on fatigue. S.T.P. No.9–J. $11\times8\frac{1}{2}$ in., pp.76. Philadelphia, 1960: The Society. (Price $3.50.)

AMERICAN SOCIETY FOR TESTING MATERIALS. '*Manual on Industrial Water and Industrial Waste Water*'. 2nd ed. Prepared by ASTM Committee D–19 on industrial water. S.T.P. No.148–D. 9×6 in., pp.xii+653. Illustrated. Philadelphia, 1960: The Society. (Price, Members $8.80; Non-members $11.00.)

APPLEGATE, L. M. '*Cathodic Protection*'. 9×6 in., pp.ix+229. Illustrated. New York; London, 1960: McGraw Hill Book Co. Inc. (Price £3 10*s*.)

THE AUSTRALASIAN INSTITUTE OF MINING AND METALLURGY. '*Sintering Symposium, Port Pirie, South Australia, September 1958*'. $8\frac{1}{2}\times5\frac{1}{4}$ in., pp.vii+381. Illustrated. Melbourne, 1958: The Institute. (Price, Members £A1.10*s*; Non-members £A2 10*s*.)

BASHFORTH, G. R. '*The Manufacture of Iron and Steel*'. Vol. 3. Steelworks fuels, furnaces, refractories, and instruments. $8\frac{1}{4}\times5\frac{1}{2}$ in., pp.viii+246. Illustrated. London, 1960: Chapman & Hall Ltd. (Price £1 15*s*.)

BOOTH, A. D. (Editor.) '*Progress in Automation*'. Vol. I. $8\frac{1}{2}\times5\frac{1}{2}$ in., pp.viii+231. Illustrated. London, 1960: Butterworths Scientific Publications. (Price £2 2*s*.)

BRITISH IRON AND STEEL RESEARCH ASSOCIATION (BISRA). '*Annual Report*' 1959. $9\frac{1}{2}\times6$ in., pp.159. Illustrated. London, 1960: The Association.

BRITISH STANDARDS INSTITUTION. '*British Standards Yearbook*' 1960. $8\frac{1}{4}\times6$ in., pp.592. London, 1960: The Institution. (Price 15*s*.)

BRITISH STANDARDS INSTITUTION. '*The sampling of Coal and Coke*'. Part I. Sampling of Coal. BS.1017: Part 1: 1960. $8\frac{1}{4}\times6$ in., pp.124. Illustrated. London, 1960: The Institution. (Price £1 5*s*.)

BRITISH STANDARDS INSTITUTION. '*The sampling of coal and coke*'. Part 2. Sampling of coke. BS.1017: Part 2: 1960, $8\frac{1}{4}\times6$ in., pp.106. Illustrated. London, 1960: The Institution. (Price £1 5*s*.)

BRITISH STANDARDS INSTITUTION. '*Methods for the Analysis and Testing of Coal and Coke*'. Part 15. Fusibility of Coal Ash and Coke Ash. BS.1016: Part 15: 1960. $8\frac{1}{4}\times6$ in., pp.13. London, 1960: The Institution. (Price 4*s*. 6*d*.)

BRITISH STANDARDS INSTITUTION. '*Specification for Thermographs (Bimetallic Type)*'. BS.3231: 1960. $8\frac{1}{4}\times6$ in., pp.7. London, 1960: The Institution. (Price 3*s*. 0*d*.)

CANNON, C. G. (Editor.) '*Electronics for Spectroscopists*'. By members of the Photoelectric Spectrometry Group and the Electronics Department of Southampton University. Schwarz Memorial Volume. $8\frac{1}{4}\times5\frac{1}{2}$ in., pp.[iv]+333. Illustrated. London, 1960: Hilger & Watts Ltd. (Price £3.)

CENTRE D'ÉTUDES SUPÉRIEURES DE LA SIDÉRURGIE (CESSID). R. Ferry '*Coulée de l'Acier*'. Complément au fascicule du Cahier XV. '*Les Gros Lingots de Forge Coulée-Structure*'. Les Cahiers du Cessid–XV. No.50. $10\frac{1}{4}\times8\frac{1}{4}$ in., loose leaf, pp.41. Illustrated. Metz, 1959: CESSID. (Price 10NF; outside France 10.50NF.)

CENTRE D'ÉTUDES SUPÉRIEURES DE LA SIDÉRURGIE (CESSID). C. G. Thibaut. '*Qualités des Cokes Métallurgique*'. Complément au fascicule du Cahier XII. '*Matières Premières en Sidérurgie*'. Les Cahiers du Cessid–XII. $10\frac{1}{4}\times8\frac{1}{4}$ in., loose leaf, pp.38. Not continuously paginated. Illustrated. Metz, 1959: CESSID. (Price 8 NF; outside France 8.50NF.)

CENTRE D'ÉTUDES SUPÉRIEURES DE LA SIDÉRURGIE (CESSID). J. Michard. '*Étude du haut fourneau*'. Bilan thermique et échanges de chaleur. Les Cahiers du Cessid–XVII. Cahier II. No.49. $10\frac{1}{4}\times8\frac{1}{4}$ in., loose leaf, pp.127. Illustrated. Metz, 1959: CESSID. (Price 30NF; outside France 32NF.)

CENTRE D'ÉTUDES SUPÉRIEURES DE LA SIDÉRURGIE (CESSID). P. Vallet. '*Chimie Physique Appliquée a la Sidérurgie*'. Réimpression. Les Cahiers du Cessid. $10\frac{1}{4}\times8\frac{1}{4}$ in., loose leaf, pp.245. Illustrated. Metz, 1959: CESSID. (Price 35NF; outside France 37.50NF.)

CENTRE D'ÉTUDES SUPÉRIEURES DE LA SIDÉRURGIE (CESSID). J. Ulmo. '*Éléments de Statistique*'. Réimpression. Les Cahiers du Cessid. No.31. $10\frac{1}{4}\times8\frac{1}{4}$ in., loose leaf, pp.140. Illustrated. Metz, 1959: CESSID. (Price 25NF; outside France 27.50NF.)

CRANDALL, S. H. (Editor.) '*Random Vibration*'. [Contributions by S. H. Crandall, I. Dyer, R. M. Mains, F. A. McClintock, K. J. Metzgar, T. H. Pian, H. Poritsky, A. Powell, D. E. Priest, T. P. Rona, W. M. Siebert.] Technology Press Books in Science and Engineering, 9×6 in., pp.x+423. Illustrated. New York; London: 1958: Technology Press of M.I.T., John Wiley & Sons Inc.; Chapman & Hall Ltd. (Price £3 8*s*.)

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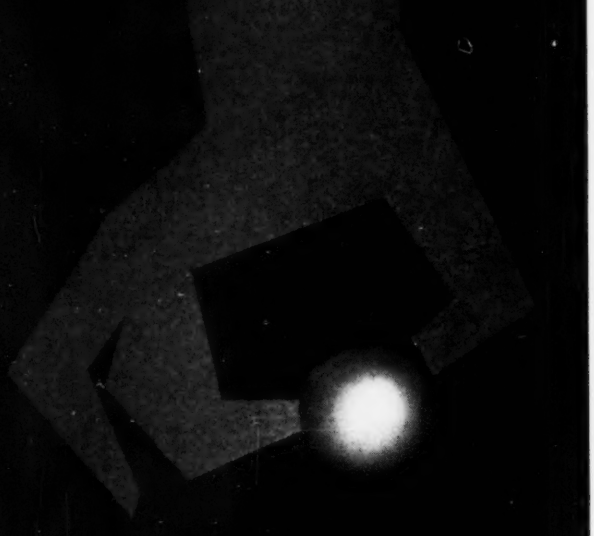

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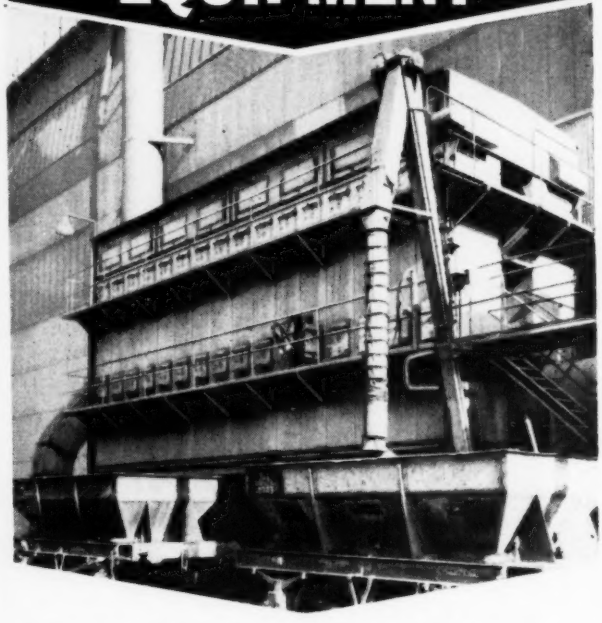

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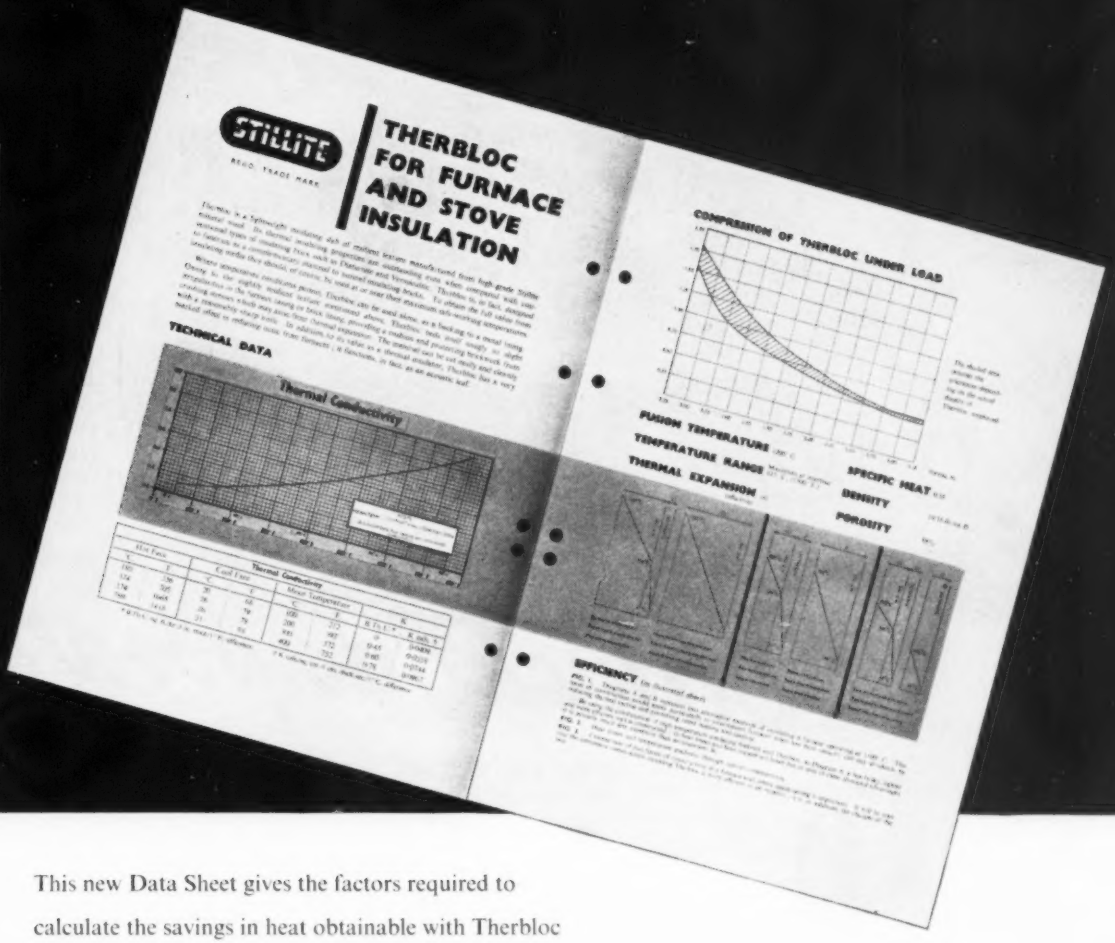

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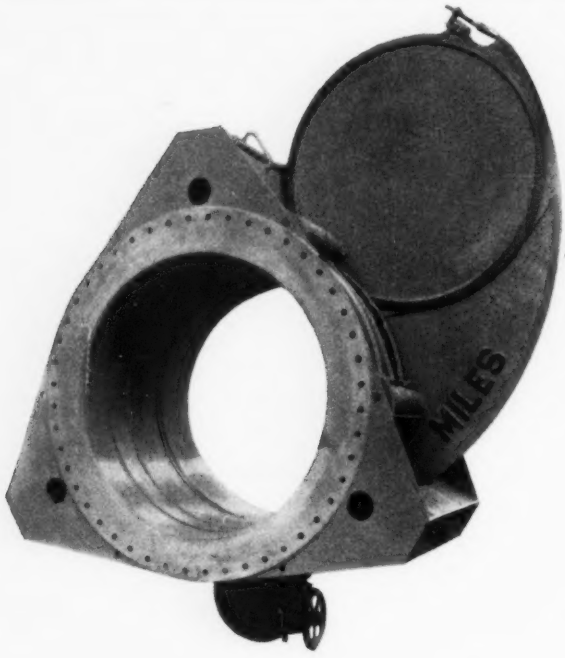

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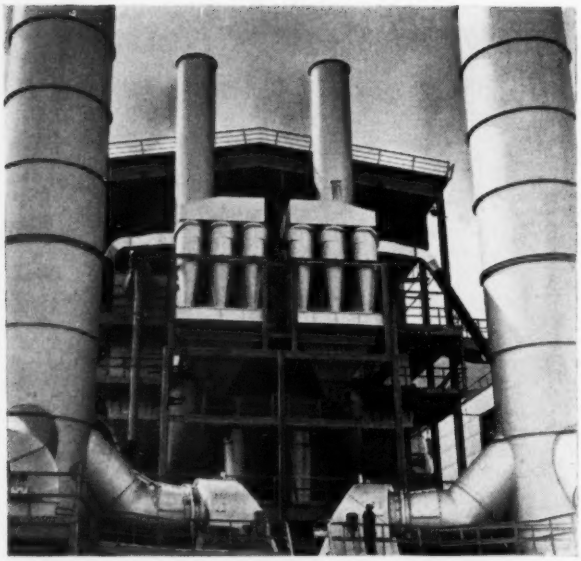

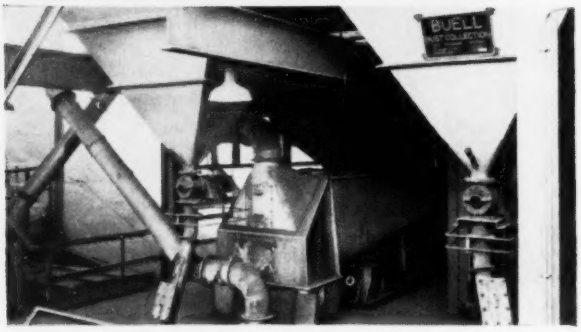

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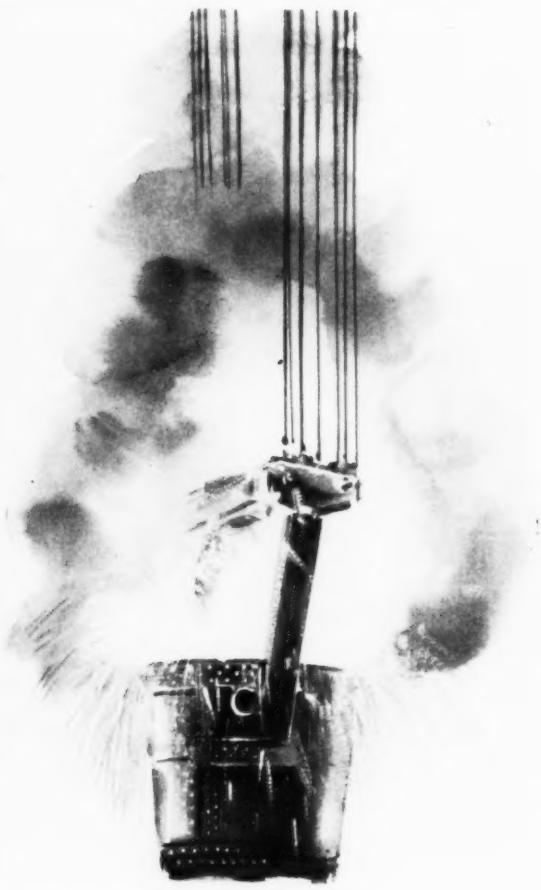

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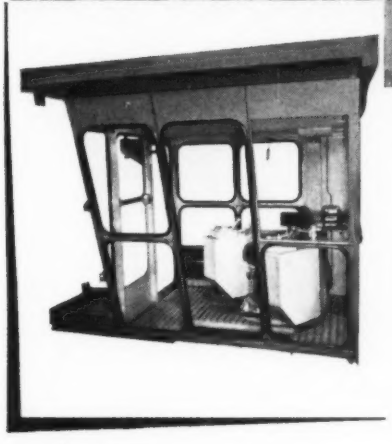

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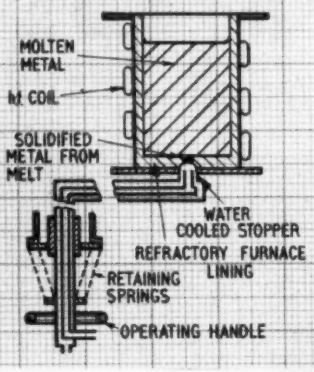

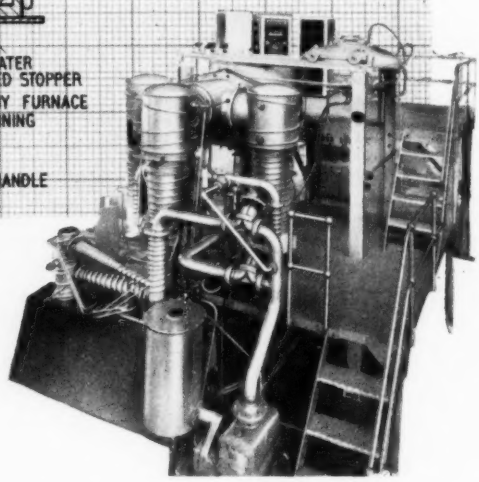

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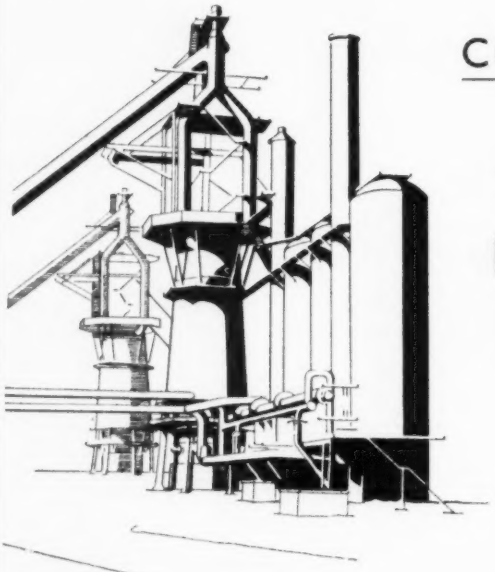

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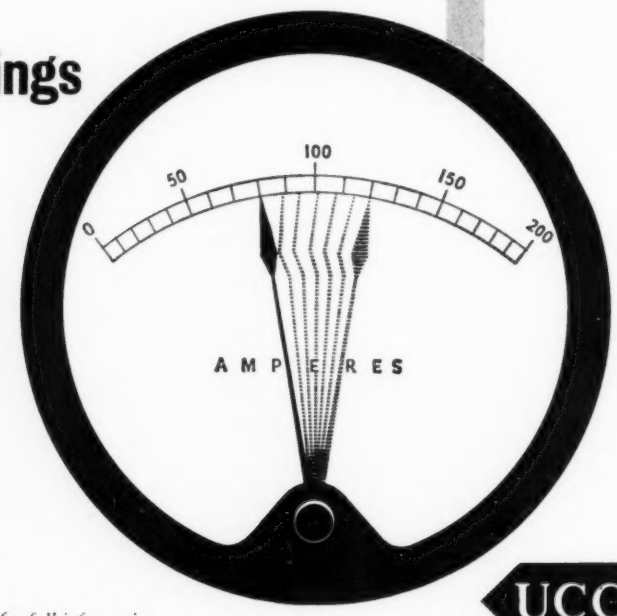

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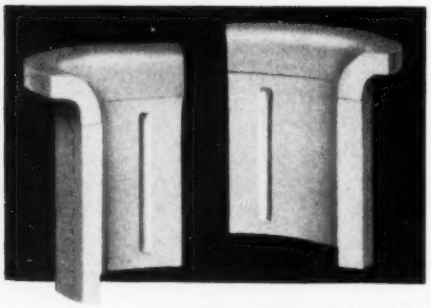

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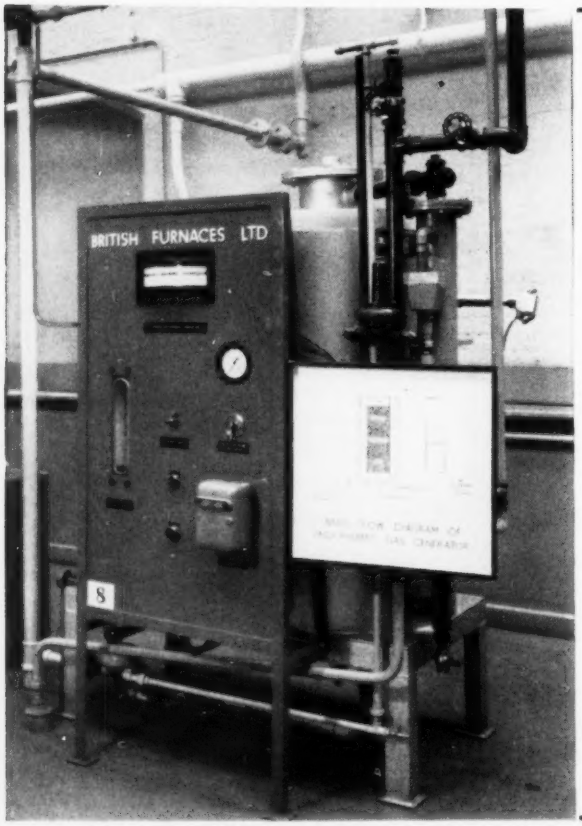

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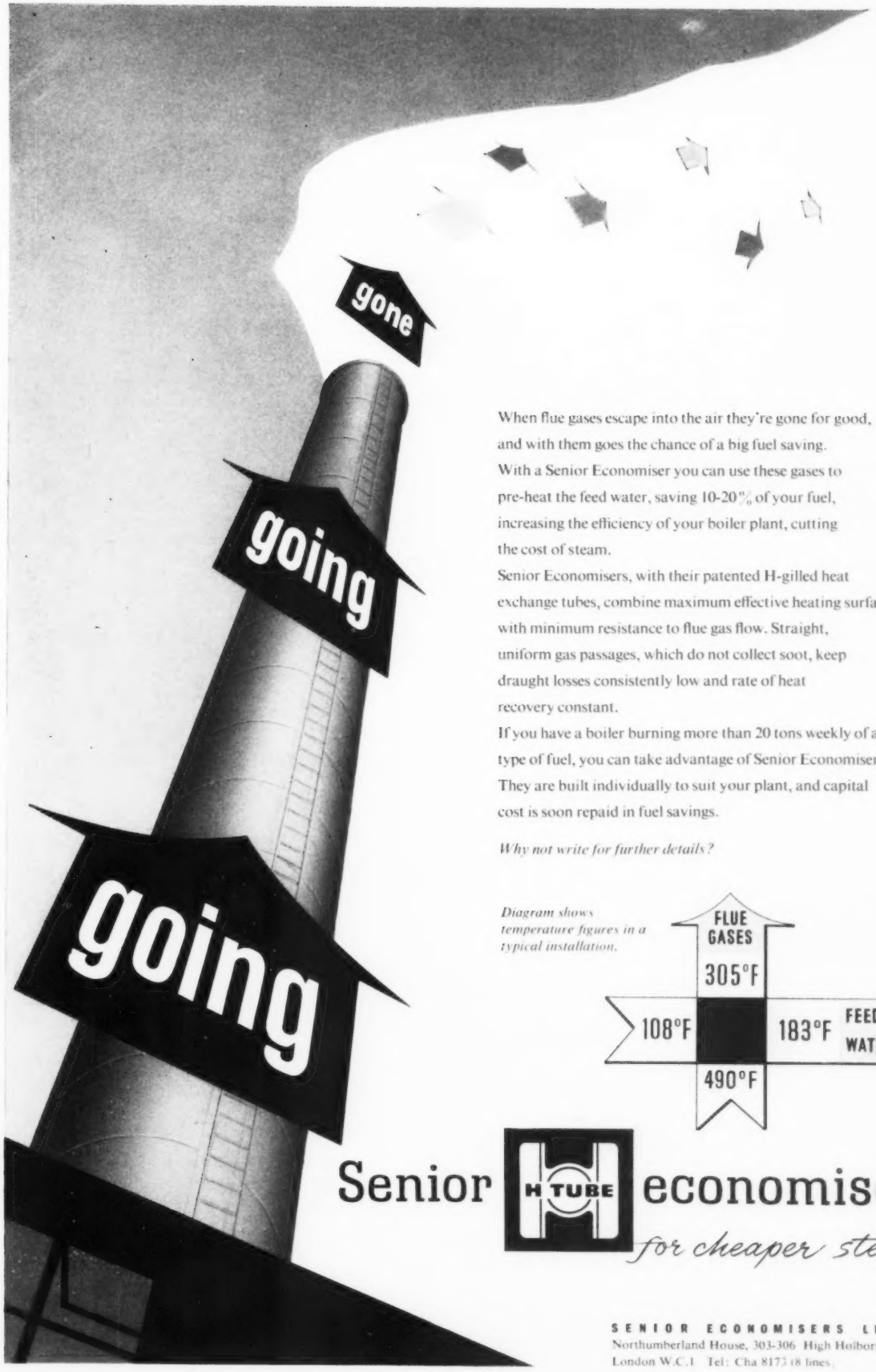

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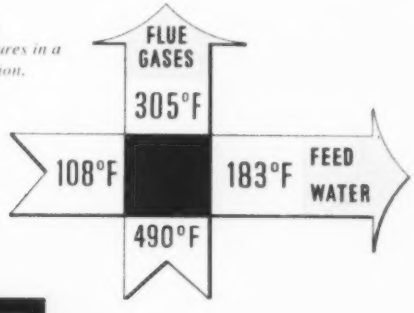

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Index to Advertisers

A.B.G. Division Cybermécanique . . —
A.C.E.C. . . . . —
Adamson-Alliance Co. Ltd. . . 2
Adamson, Joseph, and Co. Ltd. . 83
Air Products (Gt. Britain) Ltd. . 60
Allen West & Co. Ltd. . . . —
Appleby Frodingham Steel Co. . . 5
Armstrong Whitworth (Metal Industries) Ltd. 27
Arrol, Sir Wm., & Co. Ltd. . . —
Ashmore, Benson, Pease, & Co. Inside Back Cover
Associated Electrical Industries Ltd. 32

Baker Platinum Division . . . . —
Barrow Ironworks Ltd. . . . —
Beardmore, William, & Co., Ltd. . —
Bonnybridge Silica & Fireclay Co., Ltd. . . . . —
Bradley & Foster Ltd. . . . —
Brightside Foundry & Engineering Co., Ltd., The 35
British Acheson Graphite Ltd. . . 12
British Electrical Development Association . . . . . —
British Furnaces, Ltd. . . . 86
British Iron & Steel Federation, The —
British Paints, Ltd. . . . . —
British Rollmakers Corporation, Ltd. —
British Timken: Division of the Timken Roller Bearing Co. . . 63
Brookhirst-Igranic Ltd. . . . 59
Brown Bayley Steels, Ltd.. . . 44
Buell Ltd. 82
Burn Fireclay Co. Ltd., The . . —
Bushing Co. Ltd., The . . . 39

Carblox Ltd. . . . . —
Chemical Construction (G.B.) Ltd. . 58
Colville Group of Companies, The . 64
Consett Iron Co. Ltd. . . . 25
Contactor Switchgear, Ltd. . . 85
Coppee Co. (Gt. Britain) Ltd., The . 47
Cowans Sheldon & Co., Ltd. . . 62
Craven Bros. (Crane Division), Ltd. . —
Crooks, L. E. 78

D.M.M. Machinery, Ltd. 9, 61
Davy & United Engineering Co. Ltd. 72
Demag . . . . . —
Derbyshire Stone Quarries Ltd. . —
Dewhurst & Partner, Ltd.. . . —
Distington Engineering Co. Ltd. —
Dollery & Palmer Ltd. . . . —
Dyson, J. & J., Ltd. Inside Front Cover,

Edwards High Vacuum Ltd. . . . 84
Electric Resistance Furnace Co. Ltd. —
Electroflo Meters Ltd. Co. —
Engelhard Industries Ltd. . . —
English Electric Co. Ltd., The . 8
English Steel Rolling Mills Corporation, Ltd. 41

Foster, Henry, & Co. Ltd.. . . . —
Foster Instrument Co. Ltd. . . —
Fox, Samuel, &. Co Ltd. . . . 4

G.W.B. Furnaces, Ltd. 55
General Electric Co. Ltd., The . . 22
General Refractories, Ltd. . . . 13
Gibbons Brothers, Ltd. . . . —
Gibbons (Dudley) Ltd. . . . 14
Great Lakes Carbon International Ltd. . . . . —
Guest, Keen, Iron & Steel Co. Ltd. . 76
Guest, Keen & Nettlefolds, Ltd. . 38

Hadfields, Ltd. . . . . —
Head, Wrightson, Iron & Steel Works Engineering, Ltd. . . . . —
High Speed Steel Alloys, Ltd. . . —
Holmes, W. C. & Co., Ltd. . . 53
Holset Engineering Co., Ltd., The . 20
Honeywell-Controls, Ltd. . . . 73
Howden, James & Co., Ltd. . . —
Hudswell Clarke & Co. Ltd. . . 23
Hunslet Engine Co. Ltd. . . . —
Huntington, Heberlein & Co. Ltd. . —

Ilford, Limited 16
Industrial & Metallurgical Equipment Ltd. 80
International Combustion Products Ltd. 11
International Construction Co. Ltd., The 45

Johnson's Rolls Ltd. . . . . —
Johnson Matthey & Co. Ltd. . . 18

Kelvin Hughes Ltd. . . . . —
Kestner Evaporator & Engineering Co. Ltd. 80
Kingscliffe Insulating Products, Ltd. —

Lamberton & Co. Ltd. Outside Back Cover
Lanarkshire Steel Co. Ltd., The . —
Lancashire Steel Manufacturing Co. Ltd. 54
Laurence, Scott & Electrometors Ltd. —
Lodge-Cottrell Ltd. 24
Loewy Engineering Co. Ltd., The . 10
Lysaght's John, Scunthorpe Works, Ltd. 33

Marshall, Thomas, & Co. (Loxley) Ltd . 34
Metallurgical Engineers Ltd. . . —
Midland Silicones Ltd. . . . —
Mitchell Engineering Ltd. . . —
Mobil Oil Co. Ltd. 26
Mond Nickel Co. Ltd. . . . 50
Moorwood, John M., Ltd. . . . —
Morgan Refractories Ltd. . . . 81
Morgardshammars Mekaniska Verkstads AB . . . . . —
Moxey Ltd. . . . . —

National Industrial Fuel Efficiency Service . . . . . —
Newell, Ernest, & Co. Ltd. . . —
Newton, Chambers & Co. Ltd. . . —

Oughtibridge Silica Firebrick Co. Ltd. 36, 37

Pantin, W. & C., Ltd. 40
Park Gate Iron & Steel Co. Ltd. . —
Parsons, C. A. & Co. Ltd. . . . —
Pearson, E. J. & J., Ltd. . . . 82
Pickford Holland & Co. Ltd. . . 46
Power-Gas Corporation Ltd., The . —
Power Petroleum Co. Ltd., The . —
Press, William, & Son Ltd. . . —
Priest Furnaces Ltd. 77

RN Process Agency . . . . . —
Reavell-Fahie Ltd. 21
Renold Chains Ltd. . . . . —
Rawlplug Co. Ltd., The . . . 28
Research & Control Instruments Ltd. —
Richardsons Westgarth & Co. Ltd. . —
Robertson, W. H. A. & Co., Ltd. . 67
Rust-Oleum Corporation . . . 51

Salem Engineering Co. Ltd. . . . —
Schloemann AG 65
Senior Economisers Ltd. . . . 87
Sheepbridge Equipment Ltd. . . 29
Shell-Mex & B.P., Ltd. . . 42, 43
Simon-Carves Ltd. 19
Skefko Ball Bearings Co. Ltd., The . —
Smith, John (Keighley), Ltd. . . —
Smith & McLean Ltd. . . . 56
Smith, Thomas & Sons (Rodley) Ltd. —
Spencer-Bonecourt-Clarkson Ltd. . —
Stanton Ironworks Co. Ltd., The . —
Stedall & Co. Ltd 51
Steel Co. of Wales, The . . . 30
Steel, Peech & Tozer Ltd. . . . 6
Stein & Atkinson, Ltd. . . . —
Stein, John G. & Co. Ltd. . . . —
Stephens Silica Brick Co. Ltd. . . 74
Stewarts & Lloyds, Ltd. . . . —
Stillite Products Ltd. 79
Strachan & Henshaw Ltd. . . . —
Sturtevant Engineering Co. Ltd. . 17
Summers, John & Sons, Ltd. . . —
Summerson, Thos. & Sons, Ltd. . 75
Swann, Ratcliffe & Co. (Brassington) Ltd. . . . . —

Tattersall & Son Ltd. 84
Thermal Efficiency Ltd. . . . —
Thermal Syndicate Ltd. . . . —
Thomas (Richard) & Baldwins Ltd. . —
Thornton, B., Ltd. 49

Union Carbide Ltd. . . . . —
United Coke & Chemicals Co. Ltd. . 86
United Steel Companies Ltd. 3, 4, 5, 6, 7

Vaughan, Edgar, & Co. Ltd. . . . —
Visco Engineering Co. Ltd. . . 31

Ward, Thos. W., Ltd. 57
Warner & Co. Ltd. 15
Wellman Bibby Co. Ltd. . . . —
Wellman Smith Owen Engineering Corporation Ltd., The . . . 1
West Hunwick Silica & Firebrick Co. Ltd. . . . . —
West Instruments Ltd. . . . 78
Whitehead Iron & Steel Co. Ltd. . 52
Wild Barfield Electric Furnaces Ltd. —
Willan, G. L. Ltd. 48
Wincott, G. P. Ltd. 66
Woodall-Duckham Construction Co. Ltd. 70
Workington Iron & Steel Co. . . 3
Wragg, Thos., & Sons Ltd. . . 68

Yorkshire Engine Co. Ltd. . . . —

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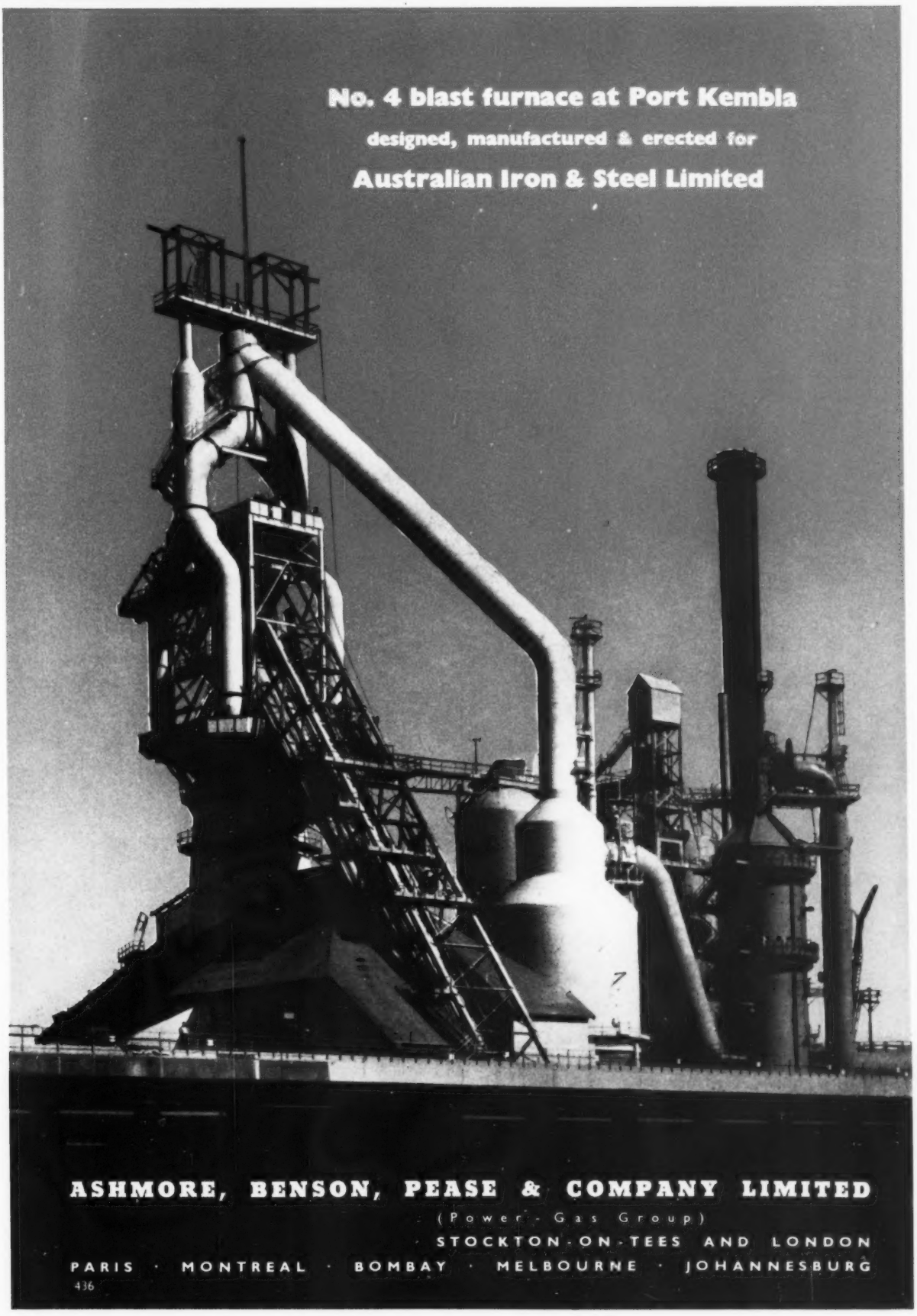
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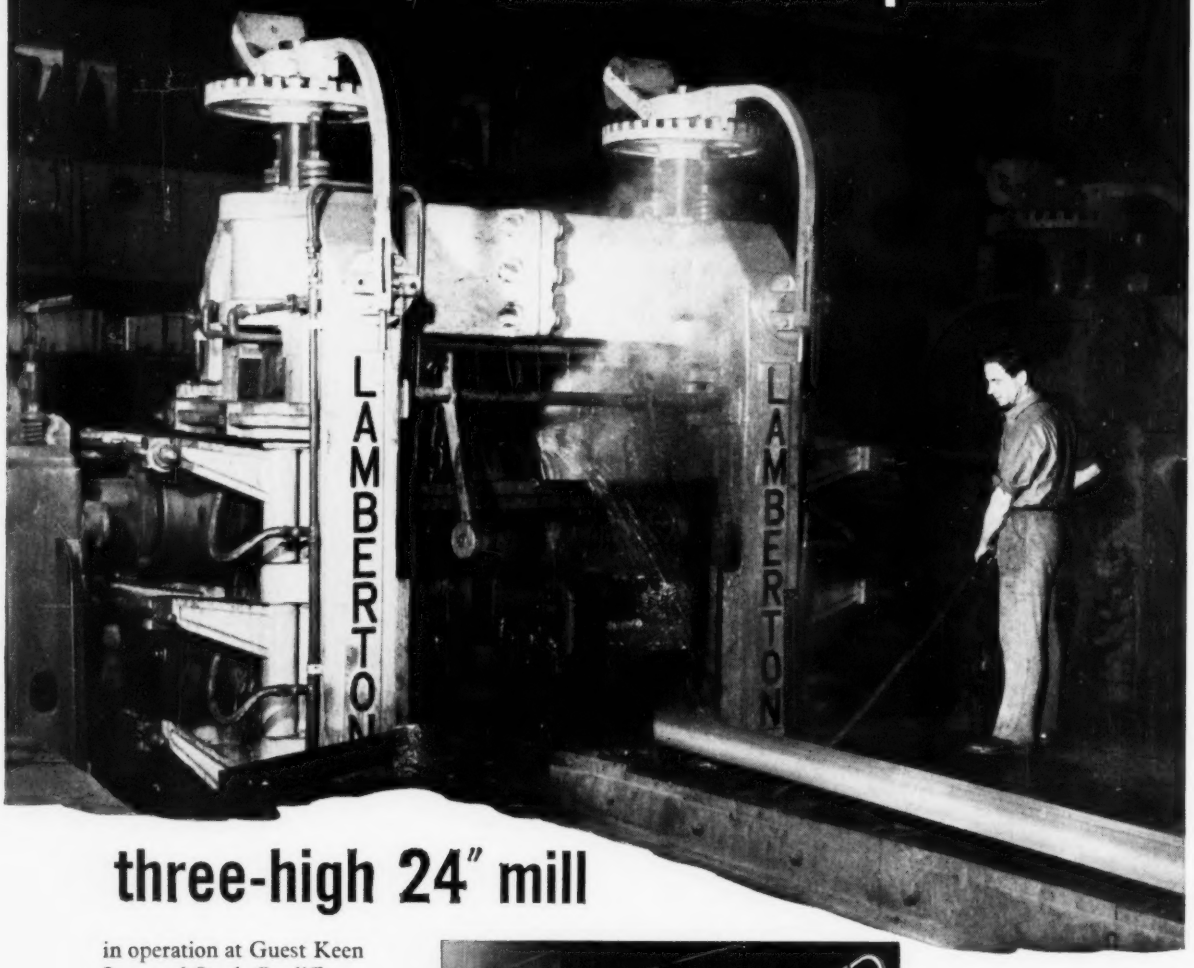
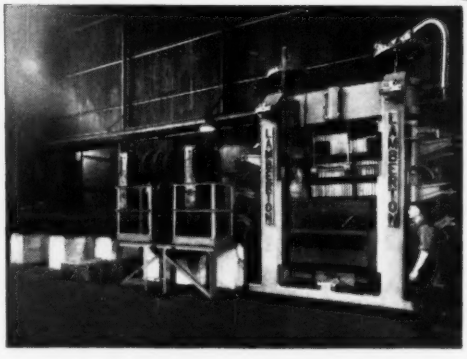


three-high 24″ mill

in operation at Guest Keen Iron and Steel, Cardiff. This mill has solid one-piece housings giving stiffness and minimum stretch. This and other features help to hold the size of the finished product constant and accurate. The stand is removed complete and in the Roll Shop the rolls are changed through the window by a special rig (see inset)



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